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METALS & ALLOYS

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The Magazine of Metallurgical Engineering

MAY 13 1932

INCLUDING

CURRENT METALLURGICAL ABSTRACTS

DOLLAR LINER "SS" PRESIDENT COOLIDGE (*Courtesy Westinghouse Electric & Manufacturing Company*)

VOLUME 3 — NUMBER 5

MAY 1932

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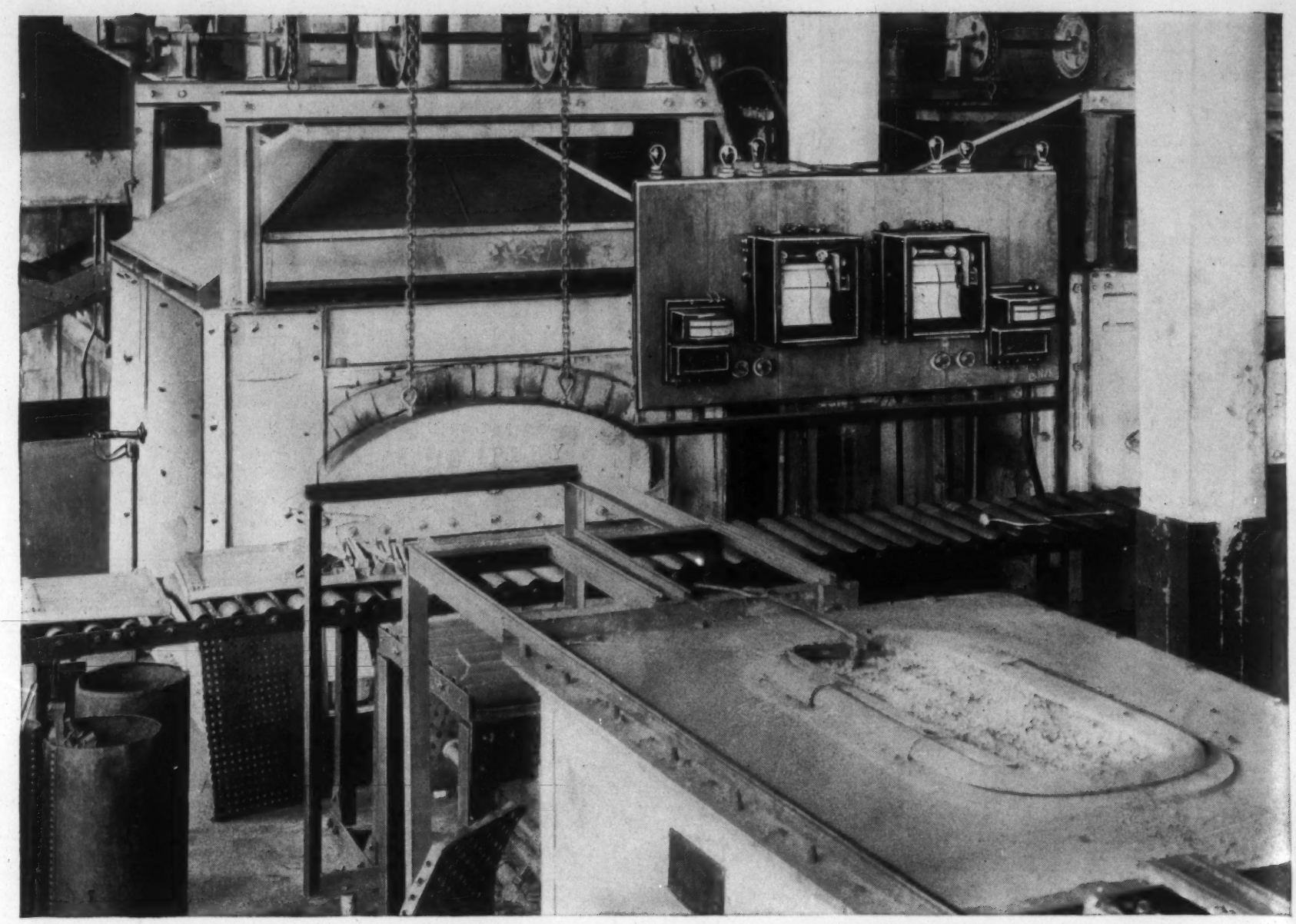

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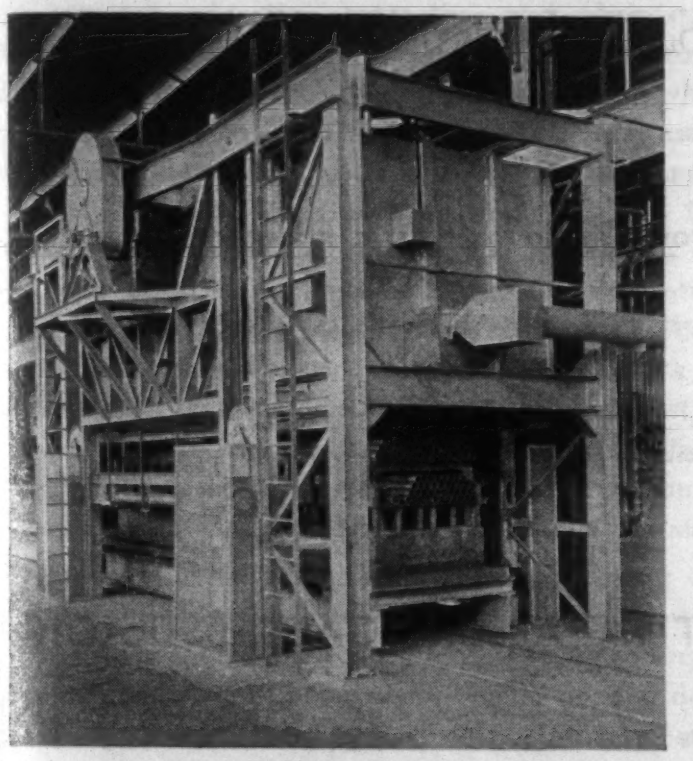

G-E elevator-type annealing furnace with controlled atmosphere and recirculating cooling system. Note charge of tubes.

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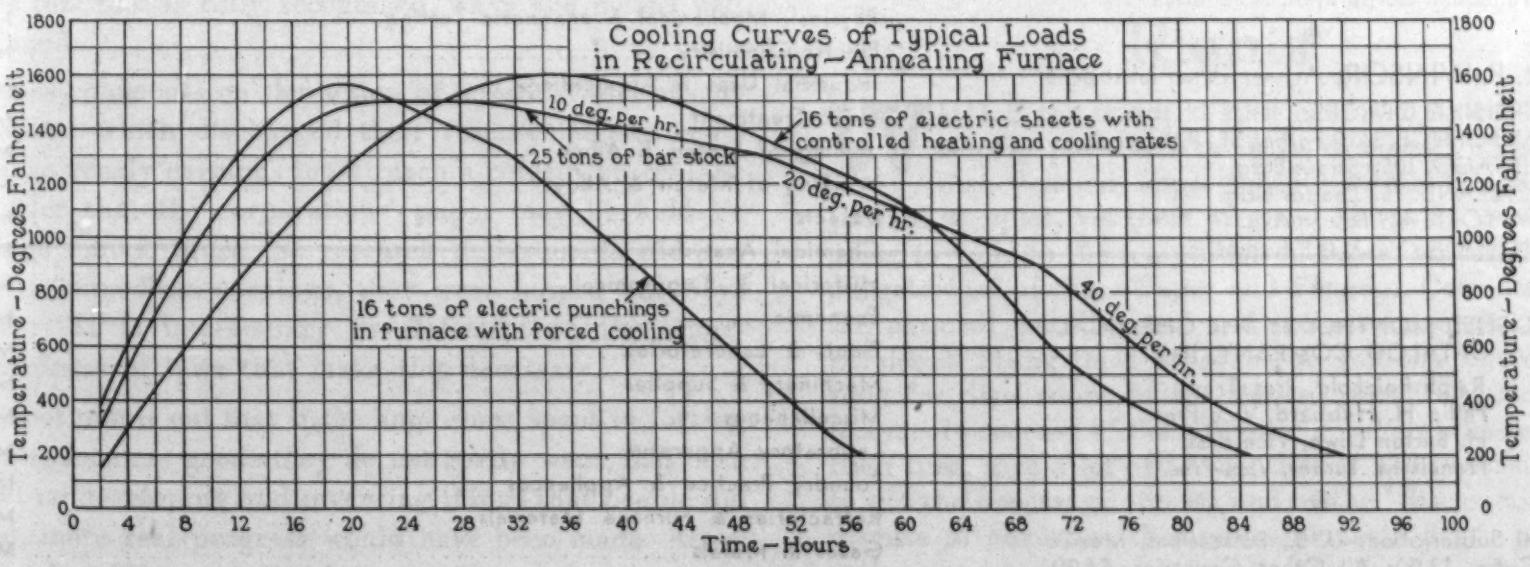
Cooling Curves of Typical Loads
in Recirculating—Annealing Furnace
Temperature—Degrees Fahrenheit
Time—Hours
10 deg. per hr.
25 tons of bar stock
16 tons of electric sheets with controlled heating and cooling rates
20 deg. per hr.
40 deg. per hr.
16 tons of electric punchings in furnace with forced cooling
0 200 400 600 800 1000 1200 1400 1600 1800
0 4 8 12 16 20 24 28 32 36 40 44 48 52 56 60 64 68 72 76 80 84 88 92 96 100

570-181

GENERAL ELECTRIC

METALS & ALLOYS

The Magazine of Metallurgical Engineering

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CONTENTS—MAY, 1932

Bright Annealing Furnace Economies 113
H. Hines

Arsenic in Ternary Lead Base Bearing Alloys . . . 116
K. H. Wegner

Hardness of Duralumin with Time and at Different Depths 120
Franz Bollenrath

Relation of Iron-Nitride to Plastic Flow and to the Fry Etch 127
H. W. Gillett (Extended Abstract)

Editorial Comment 111
Review of Manufacturers Literature . . . . . MA 156

CURRENT METALLURGICAL ABSTRACTS

General ..... MA 115
Properties of Metals ..... MA 116
Properties of Non-Ferrous Alloys ..... MA 117
Properties of Ferrous Alloys ..... MA 118
Corrosion, Erosion, Oxidation, Passivity & Protection of Metals & Alloys . MA 119
Structure of Metals & Alloys ..... MA 121
Physical, Mechanical & Magnetic Testing ..... MA 127
Electro-Chemistry ..... MA 129
Industrial Uses & Applications ..... MA 130
Heat Treatment ..... MA 131
Joining of Metals & Alloys ..... MA 133
Working of Metals & Alloys ..... MA 135
Defects ..... MA 140
Chemical Analysis ..... MA 140
Historical & Biographical ..... MA 141
Economic ..... MA 141
Plants & Laboratories ..... MA 142
Machinery & Supplies ..... MA 142
Miscellaneous ..... MA 143
Laboratory Apparatus ..... MA 143
Foundry Practice & Appliances ..... MA 144
Refractories & Furnace Materials ..... MA 146
Gases in Metals ..... MA 146
Inspection ..... MA 147
Effects of Elements on Metals & Alloys ..... MA 147
Instruments & Controllers ..... MA 148
Effect of Temperature on Metals & Alloys ..... MA 148
Reduction Metallurgy ..... MA 152
Non-Metallic Coatings for Metals & Alloys ..... MA 153
Ore Concentration ..... MA 154

EDITORIAL COMMENT

Research In Times of Depression

TWO recent articles afford interesting points of view on the subject of research in times of depression. These are "Progress and Prosperity: A Suggested Program" by Professor Adelbert Ames, Jr.,[1] of Dartmouth, and "The Place of Research in Industry," by Dr. F. B. Jewett,[2] head of the Bell Telephone Research Laboratories.

The first is a thoughtful analysis of the true economic wealth and degree of civilization of a nation, with emphasis on the importance of "non-consumable intangibles." Among these he lists, in this order, religion, arts and music, research, education, recreation, and administration. While the monetary profits that naturally result from putting capital into research tend to promote it, and while the wisdom of certain men of wealth has led them to establish Research Foundations, yet recognition of the benefits to be derived should lead, in Ames' mind, to still further support of research, among the other intangibles of this class, by taxation for such activities, by exempting all such activities themselves from taxation by revision of laws and business policies to give greater profits on investments in these fields, and by promoting the establishment of foundations for carrying on these activities, such foundations to be supported by joint contributions from those not rich enough to set up individual foundations.

Chief emphasis is laid upon research, among the desirable non-consumable intangibles, because "new developments are the life of economic prosperity." Surplus of capital and labor in boom times should go into development rather than into replacements. Replacements may keep economic well-being in *status quo,* but do not increase it. Greater economic wealth will, in the long run, result from diverting capital to the field of development. Once this fact is fully recognized, ways and means will be found.

Ames comments on the action of several railroad corporations which discharged their research departments because yearly earnings must reach a certain percentage in order that the corporations' paper may be held by savings banks. Since the research departments did not add to immediate earnings, they were lopped off. Says Ames, "It is unnecessary to elaborate on the shortsightedness of laws that make this necessary."

Cabot points out that if the huge sums spent in forcing upon consumers goods they do not really want, had been spent on developing and inventing things that people did want, more real progress would have been made. Ames' plan of making new discoveries and inventions a prime objective seems sound to Cabot.

Jewett's article is written from the point of view of an industry that itself long ago came to the same general conclusions that Ames expresses. He says, "The research laboratory is the place where progress can be made most rapidly, most cheaply and with the greatest assurance of avoiding missteps." This is true not only in good times, "but also in periods of depression when the problems are those of doing existing things in more economical fashion, or of developing new things for which a demand can be created. The research laboratory has proven itself time and again to be the saving factor in what might otherwise have been a very unpleasant situation." His laboratories are therefore retaining their trained scientific personnel without any cuts.

Jewett states that his own organization, working in a non-competitive field, has many research projects that could just as well be handled coöperatively, were there other large competing units in their specialized field. Since only very large industrial units can maintain wholly sufficient and self-contained research laboratories, the smaller units are placed at a great and frequently impossible disadvantage unless they can find an alternative way to place themselves on a plane of equality in scientific knowledge and control. Common effort in the acquisition of basic knowledge, supplemented by individual effort in its adaptation to the needs of an individual firm, is the logical method of procedure.

Both articles deserve careful study by executives who are debating whether they can continue to "afford" research in a period of depression.—H. W. GILLETT

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The Audience Has Some Rights

WHILE the editorial staff can, and should be, decidedly plain spoken in its comments on matters that demand improvement, within the field its journal occupies, yet there are some limits. So we are not going to yield to the temptation to set up an "Index Expurgatorius" of those Paper and Program Committees of the national metallurgical and technical societies which persist in filling their programs so full that there is no time for discussion; of the listless presiding officers at technical sessions; of the boresome speakers who run over their time, give a lot of useless details and generally fail to get their message across; and of the "discussers" who discuss to advertise themselves or their firm's product rather than to add to the subject in hand. In the words of the Mikado, we've "got them on the list," but we won't publish the list.

We can however do the other thing. We can make favorable comment from time to time on those Program Committees which plan sessions so that there is time for discussion; on presiding officers who catalyze discussion;

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[1] *Supplement to Dartmouth Alumni Magazine,* Vol. 24, No. 4, Jan. 1932, 15 pages. Includes comment by Prof. P. Cabot, Harvard Graduate School of Business Administration.

[2] *Proceedings, American Petroleum Institute,* Sec. III, Dec. 1931, pages 27-33.

on speakers who do leave the audience with a clear-cut impression of what it's all about; and on "discussers" who courteously point out errors and add new and pertinent facts, opinions and suggestions. There might be a sort of "All American" selection of those who thus realize that the audience at a technical meeting does have some rights.

With the aid of our Editorial Advisory Board we plan to make such comment and enter some of those who most notably deserve it, on the *Metals & Alloys* "Roll of Honor."

We'll start this by mentioning three speakers at the American Institute of Mining & Metallurgical Engineers' meeting: Dr. J. A. Gann, of the Dow Chemical Co., whose talk on magnesium at the Institute of Metals was written to include a lot of human interest, and very clearly presented; Dr. J. G. Thompson, of the Bureau of Standards, who summarized the situation on analysis for "Gases in Metals" with brevity and clarity, and Dr. R. F. Mehl, of the American Rolling Mill Co., who presented one paper and discussed others so that his hearers very definitely got the idea. One friend commented to us that we ought to get Dr. Mehl to write an article for *Metals & Alloys* on how to present a technical paper.—H. W. Gillett

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Readers' Comments

Editor, Metals & Alloys:—

The production of tungsten carbide was accomplished a few years ago, and its application to the mechanic arts during the stress of war times formed the basis of much experimental research for the further improvement of metal cutting tool alloys. Tools of this ultra-hard character are now in regular use in some highly specialized industries where rapid production is demanded in accurate machining of hard material. The results secured by the use of these diamond-like tools are truly astonishing. Information in regard to tool performance is widely available. The object of this letter is to comment on their production rather than their use.

Tools of this kind are not yet so well adapted to large work of the rougher class where deep cuts and a coarse feed with a high speed steel tool are best suited to the job, and their employment is not by any means general in common machine shop practice. Their use in many mechanical operations would be greatly extended, were it not for the difficulty of producing such tools and consequently their very high cost.

The writer has tried many experiments along this line and finally adopted the procedure embodied in the apparatus herein described, and in which the component metals of the alloy are derived from their oxides by the well-known means of alumino-thermic reduction. This is carried out however, under conditions that differ very essentially from the usual methods of practice, for the reason that the ordinary process of open air ignition for the production of this class of alloys would be resultless on account of the volatile and combustible nature of the contained elements at the extreme temperature of the reaction. The product would be consumed as fast as formed, reoxidized and lost.

In order to avoid this, some precautionary measures must be taken. The charge must have considerable mass, at least two or three pounds weight. It is useless to ignite a few ounces or any small quantity, because the heat is lost too quickly. The reaction must take place within a perfectly closed space, out of all contact with the atmosphere, and the containing vessel must have sufficient strength to withstand a very high pressure, around 2000 lbs./in.^{2} It must be provided with a heavy lining, say 1½ inches thick, of some very refractory material of a heat insulating nature, or this lining may be replaced by a thick walled covered crucible made of like material and removable at will. Under these conditions the highest possible temperature of the reaction is reached and maintained, for a sufficient length of time to effect the desired combination; that is, to form tungstic carbide alloyed with cobalt, nickel or other metals in any required proportion.

The container is a cylindrical pot shaped steel casting 10 inches deep and 8 inches in diameter inside, with an integral bottom and a wide flange on top. The wall thickness is ½ inch. The head or cover is fitted to the container flange with a perfectly tight ground joint, and is held down by a number of strong bolts. Within the container is the crucible or lining for the reception of the charge, consisting of the required oxides and powdered aluminum (and carbon if a carbide is wanted) in proportionate amounts calculated from the metallic content of the oxides and that of the specified alloy, but the weight of aluminum used is always made a little short of the exact amount required to completely reduce the oxides, and the addition of about 1% of lime to the charge seems in some way to cause an increase in the carbon content of the alloy, but this is not imperative, and in the case of non-carbon alloys is omitted.

All the materials of the charge should be finely powdered and well mixed by stirring and sifting several times repeated, but prolonged trituration is entirely unnecessary.

The ignition of the charge is started electrically by means of a specially constructed spark plug entering through the container head. When the charge is fired, there is no immediate indication of the intense heat and enormous pressure within, but after a few minutes the outside walls of the container become sensibly warm, and soon after this the relief cock may be opened and the pent-up gases discharged. Upon the removal of the head, the alloy will be found under the slag in the form of a disc accurately molded to the shape of the crucible bottom.

If the alloy is one consisting largely of tungstic carbide, the disc may be broken up under a power hammer, or, with some difficulty, by repeated blows of a heavy sledge, and the fragments can, with the proper equipment, be remelted and cast into tool shapes of any desired pattern, but the temperature required is high, and some special arrangements are necessary. In this connection one remarkable fact may be noted: viz., the addition of zirconium (itself almost infusible) in any considerable amount, say 10-15%, very materially lowers the fusion point of the resulting alloy, and this, of course, enables the subsequent melting and casting to be conducted under more manageable conditions. In any case, however, these secondary operations demand the use of the highest class refractories in the make-up of the crucibles and furnace linings. There is, however, a modification of this process by which these tool bits or shapes of the hardest possible grade may be made in situ, formed in the first instance, and without subsequent melting and casting.

118 Beaumont St. Samuel M. Howell
Zanesville, Ohio

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Chuckles

Instead of "hardened copper," this time it is "softened gold." From the New York Times of January 22, 1932, under Mexico City date line, is the statement that Professor Ramon C. Robles, pure Mixtecan Indian, federal inspector of schools for the State of Oaxaca, declared that the ancient Mixtecans, who were the finest goldsmiths ever known to Western civilization, had an art of mixing metals with juices of herbs which rendered these metals so pliable they could be moulded like clay. This, he says, is the way the present treasures found at Monte Alban were fashioned, as he declares these golden bracelets, earrings and necklaces bear no evidence of having been beaten into shape.

The secret has been zealously guarded, he said, in the hills of Oaxaca and still is practiced, but it never has been revealed, because some Mixtecans never have acknowledged the dominance of the white man.

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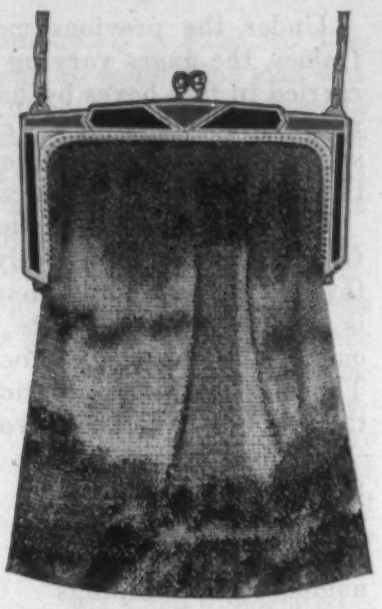
The Washington Award for 1932, "in recognition of devoted, unselfish, and preeminent service in advancing human progress," was presented to **Dr. William David Coolidge**, associate director of the research laboratory of the General Electric Company at Schenectady, following a dinner and reception at the Hotel Sherman, Chicago on Feb. 24. Dr. Coolidge is the eleventh to be so honored by the Washington Award Commission, the first presentation being to President Hoover, in 1919. Dr. Harry Woodburn Chase, president of the University of Illinois, delivered the principal address of the evening, evaluating the importance of research in the present social order.

Among the best-known contributions of Dr. Coolidge are ductile tungsten for lamp filaments, wrought tungsten for contacts and X-ray targets, the hot cathode X-ray tube, various types of X-ray generating equipment, the "C" tube for submarine detection and signaling, and the cathode ray tube.

BRIGHT ANNEALING FURNACE ECONOMIES

BY H. HINES*



Whiting and Davis Company's Quality Mesh Bags.

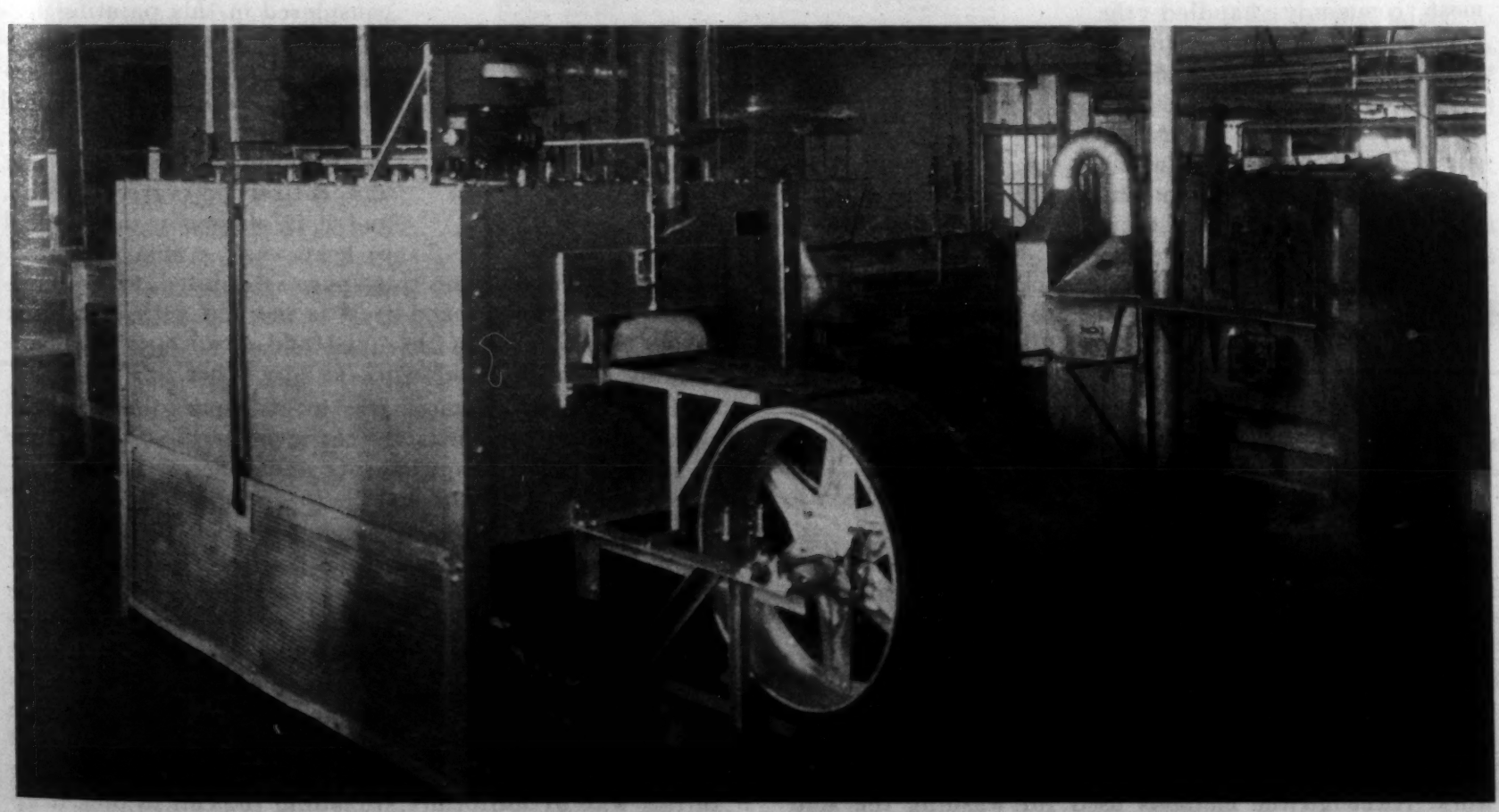
A PROPERLY selected example of improved operating technique should include not only the particular improvement which is to be stressed, but should also combine all of the latest features in the design of the equipment. The economies secured from the installation of a special atmosphere electric conveyor furnace at the plant of the Whiting and Davis Company, Plainville, Massachusetts, are the result of the combination of special atmosphere control and improved furnace design.

The Whiting and Davis Company manufacture quality metal mesh bags for ladies. These bags are made of a soldered mesh in various forms and enameled. The frames at the opening are made of plated brass or nickel silver and these frames are embossed in special designs. The embossing of the frames requires a varying number of stampings, depending upon the gage of the metal and the design used. It is necessary to anneal the frame before each stamping to secure the desired results.

Whiting and Davis installed a furnace specified by the Process Engineering and Equipment Corporation.† This was a 54 kilowatt special atmosphere electric conveyor furnace, having a heating chamber 72″ long, a cooling chamber 150″ long, a conveyor belt 12″ wide, with a raised edge and a clearance above the belt of 3″. Complete automatic temperature control equipment, conveyor belt speed regulation equipment and special atmosphere control equipment were provided.

This equipment was installed to replace box type fuel fired furnaces, having temperature indicating equipment, but hand operated burners and also to eliminate the necessity of pickling, cleaning and drying operations and equipment. The layout for the manufacture of these frames is very convenient and the fuel fired furnaces were only a short distance from the press line.

*Whiting and Davis Company.
†This company has been purchased by The Electric Furnace Company, Salem, Ohio.

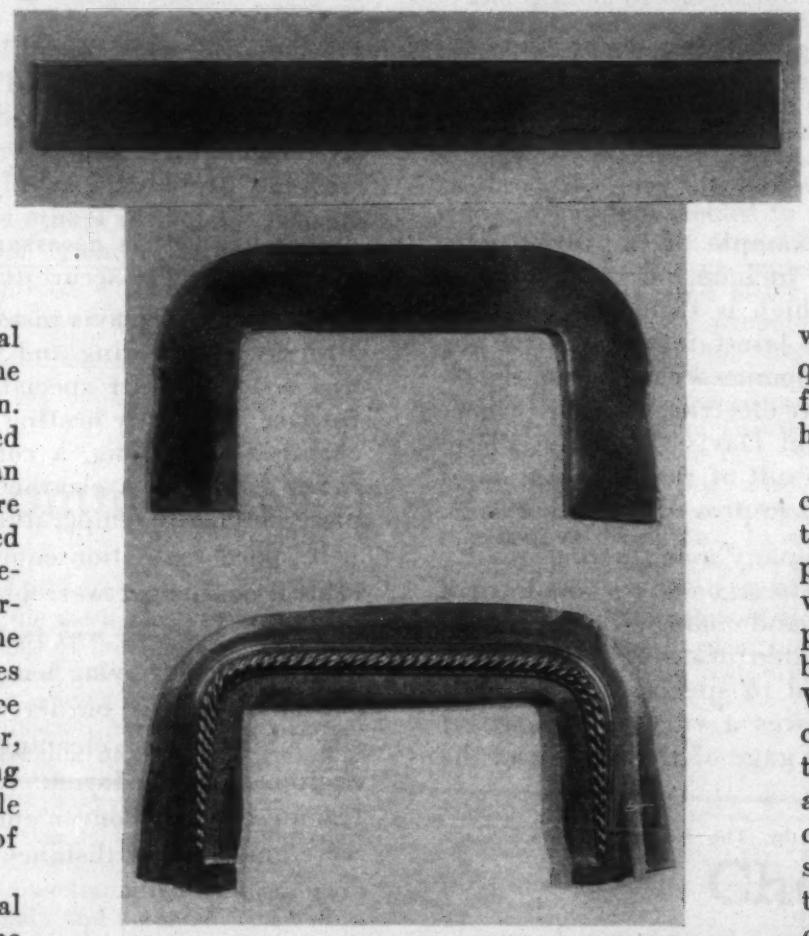


Bright Annealing Furnace. Photograph was taken before the two fuel fired furnaces were dismantled.

Under the previous method, a lot or partial lot of frames, the gages varying from 0.028 to 0.070, would be carried in tote boxes by hand truck to the annealing furnace. Nine handlings were required previous to each stamping operation before the material was returned to the tote box in an annealed and dry condition.

Under the present method, the material is transferred from the tote boxes directly to the conveyor belt, it passes through the annealing chamber and cooling chamber and is dumped down a chute into an empty tote box thoroughly and uniformly annealed and bright clean and dry. The only handling operation consists of the transfer from the tote box to the conveyor belt.

Because of the uniform quality of the anneal which resulted, it was found that the number of stampings required to bring up a given design were considerably reduced. In the previous method the number of stampings per part depended upon the quality of the anneal received by the pieces in the center of the annealing pan. This type of material packed closely and when brought to an indicated furnace temperature was never uniformly annealed in the box type furnaces. Therefore, the number of press operations were governed by the hardest material or the pieces in the center of the furnace load. In addition to this factor, the roughening of the pickling acid affected to a considerable extent the resistance to flow of the metal in the dies.



LIGHT SKELETON FRAMES

Top—Cut from Flat Stock.
Center—Bent to Shape.
Bottom—The Number of Stampings required to bring out the Design has been reduced from three to one.

In the 2 classes of material most commonly handled the stampings were reduced in one case from 6 to 4 and in the second case reduced from 3 to 2 or 1, depending upon the gage. These reductions in the number of stampings are the direct result of a more thoroughly and uniformly annealed product and the elimination of pickle etching. Therefore, as far as the annealing practice was concerned, on a frame which previously received 6 annealings and subsequently 54 handlings, in the present special atmosphere electric conveyor furnace, only 4 handlings are required. Therefore, the reduction in handlings, instead of being 9 to 1 is in the ratio of 54 to 4 or 13 to 1. This ratio in the decreased number of handlings is in the case of one type, from 27 to 1.

The elimination of 2 or more stampings for each piece fabricated also resulted in a considerable saving in press labor. This saving amounts directly to 1/3 of the direct cost of forming and stamping the frames.

One of the things noticeable, following the installation of the equipment, was the immediate elimination of the complaints on the part of the press operators because of sore hands, due to the acid which remained on the work. Although the methods used for washing the acid from the frames were in accordance with the best practice, using hot water and running cold water, a certain amount of acid remained on the pieces.

The elimination of this acid also had a beneficial effect on the surface of the dies. Because the material was more uniformly annealed and acid eliminated, it was found that by a more careful finish of the dies, a longer life of the die and improved quality resulted. The value of such a saving can not be estimated, but where expensive dies are carefully embossed, any saving would be of considerable importance.

Another saving which is closely allied with the improvement in quality is the metal surface following the last stamping. The use of pickle, the purpose of which is to clean the metal surface by removing the oxide and scale, is detrimental to the metal itself. The acid acts more rapidly along and enlarges crystal boundaries and has a tendency to generally roughen the metal surface. The pickled metal which has been stamped requires a great deal more work for polishing than that which has been bright annealed.

One additional factor is of considerable interest. At about the time that the special atmosphere electric conveyor furnace was installed, the company was painting the interior of its building with aluminum paint. Walls, ceilings, piping and machinery were painted in order to improve lighting conditions and to secure neatness and cleanliness which are highly desirable. The space surrounding the 2 fuel fired furnaces was not considered in this painting program because, even though proper hoods were provided, the annealing operation, accompanied by pickling, cleaning and drying, is a noisy, dirty, hot, wet and entirely disagreeable job, and it is obviously impractical to keep such a space painted.

At the present time this space is painted with aluminum paint. The furnace itself is painted with aluminum paint and because of the quiet and clean operation, annealing is as pleasant work as any other.

The furnace is automatic in its operation and the operator merely adjusts the conveyor or the temperature to meet standard requirements and places the work directly on the conveyor belt.

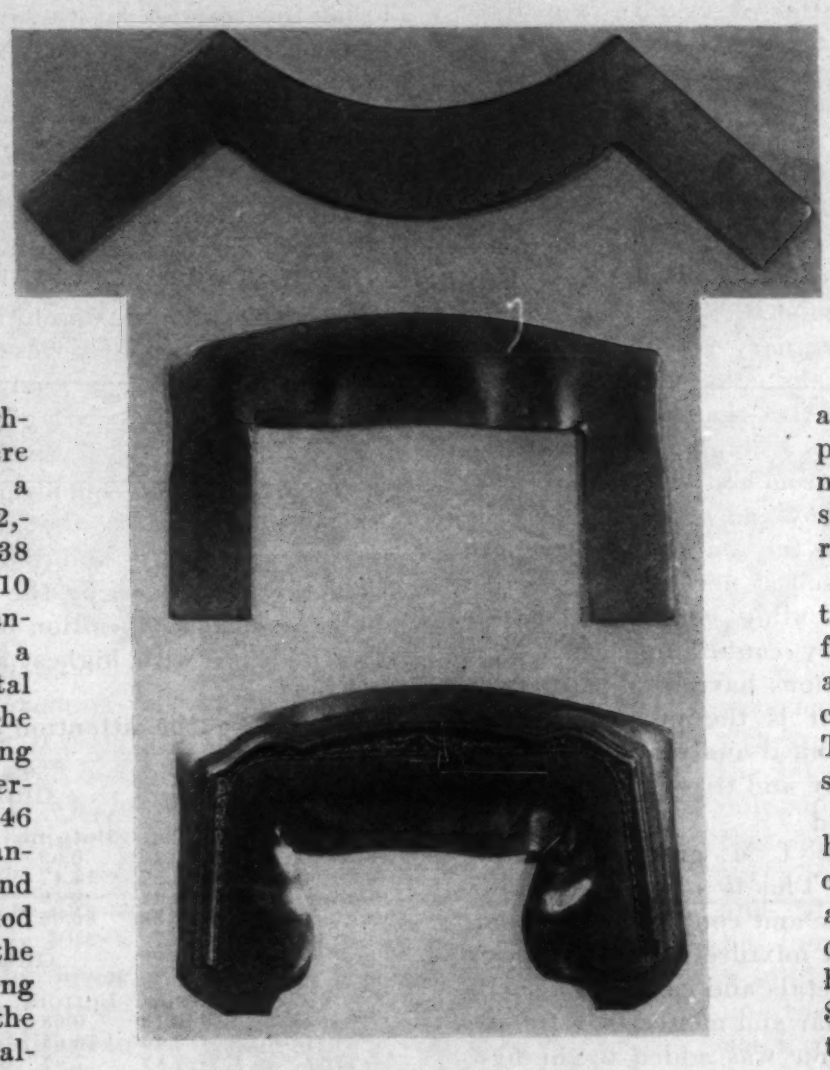
In addition to the flat stampings and shapes which are the bulk of the production, the furnace is also used for bright annealing all other materials which are part of the manufacturing requirements. The company draws a certain amount of its own brass wire and this wire is bright annealed in the conveyor furnace by spreading the coils which are 8 or 10 inches in diameter along the conveyor belt. A 10 inch coil weighing approximately 25 or 30 lbs. will be spread a distance of 2 or 3 feet. It is possible in this way by uniformly spreading the coil to pick it up again without snarling or twisting it.

Findings consisting of small cups, bent strips and so forth are annealed on the conveyor belt in fine mesh wire trays. In some cases also, although in small quantities, the company bright anneals strip sheet of various kinds. This is done by running the sheet from a coil on an arbor, the front end of the strip being attached to the conveyor belt.

The same factors of uniform anneal and the elimination of pickling, cleaning and drying operations applies with respect to the sheet, wire and findings and in some of these cases it is even more important.

To determine the savings in direct costs which resulted from the use of the special atmosphere electric conveyor furnace, 3 months' production was used as the basis. The following costs were for the 2 box type fuel oil furnaces.

In the first case 101,473 quality frames, weighing 25 lbs. per 100 pieces, were annealed 6 times each, or a weight of approximately 152,000 lbs. In addition, 493,438 skeleton frames, weighing 10 lbs. per 100 pieces were annealed 3 times each, giving a weight of 148,020 lbs. The total weight was 300,120 lbs. The number of hours of annealing operation was 869 and the average production per hour 346 lbs. The labor charged to annealing, pickling, cleaning and drying during this period amounted to $1,641.93, the average cost per hour being $1.90. During that time the number of operators for annealing, pickling, cleaning and drying varied from 3 to 5, depending upon requirements. Fuel oil for the period cost $291.08 and the cost of acid was $28.00. The total amounted to $1,961.01.



HEAVY QUALITY FRAMES

Top—Cut from Flat Stock.
Center—Bent to Shape.
Bottom—The Number of Stampings required to secure the shape and design has been reduced from six to four.

Additional expenses for the operation included water, steam, pickle baskets, air and so forth, as well as approximately 3 times the floor space used by the electric furnace. To compare with these items the electric furnace has cooling water and a small motor on its conveyor drive. Since only direct costs are considered, these smaller items are not estimated.

The cost of operating the electric furnace based on a similar production and time are as follows:

In this case 101,473 quality frames weighing 25 lbs. per 100 pieces, 4 annealings each are equivalent to 101,400 lbs. The 493,438 skeleton frames weighing 10 lbs. per 100 pieces were annealed according to standard practice, depending upon the gage. Approximately 50% were annealed once and 50% twice. The weight being equivalent to 73,910. The total weight annealed being 175,310.

The furnace was operated at an approximate production of 500 lbs. per hour, giving a total time of 351 hours. Labor for the period amounted to $245.75. The average labor cost per hour being 70 cents. One man was sufficient with the exception of 1 or 2 cases where a helper was used for moving material. The cost of power was as follows: 2600 kilowatts during heating up periods and 11302 kilowatts during hours of operation. At 2 cents per kilowatt, this amounts to $278.04.

For regular operation the furnace requires approximately 55 cubic feet of hydrogen gas and 25 cubic feet of nitrogen gas per hour. This gas can be secured either through the dissociation of liquid anhydrous ammonia or through the purchase of tank hydrogen and nitrogen. The cost per 100 cubic feet for this gas when dissociated is 50 cents, the cost for 351 hours of operation being $140.40. Therefore, the total cost of direct operation of the electric furnace amounted to $664.14, as compared with the fuel fired furnace of $1961.01. The direct saving in annealing for the period of 3 months was $1296.87.

The savings in press operation amounted to 2 stampings for each of the quality frames and 1 or 2 stampings in the case of the skeleton frames. The total number of stampings saved amounted to 943,103.

At a press rate of 700 per hour, this amounted to a saving of 1347 hours in operator's time and press time. Assuming an operator's rate at 50 cents per hour, the saving would be $673.50. This saving, added to the $1,296.87 amounts to a total saving in direct cost of $1,970.37 for a period of 3 months.

It is interesting to note that the saving in press operations is sufficient to pay the present annealing cost, so that 100% of the previous cost of annealing was saved by the installation.

◈ ◈ ◈

The offices of the **United Engineering & Foundry Co.** have been moved to the First National Bank Building, Pittsburgh, Pa.

◈ ◈ ◈

The Chicago Flexible Shaft Company announce the appointment of Mr. A. E. Goldie, 561 East 108 Street, Cleveland, as representative of Stewart industrial furnace in the Cleveland territory.

◈ ◈ ◈

Croll-Reynolds Engineering Co., Inc., New York, N. Y., has recently arranged to handle NI-Resist castings, manufactured by its associate, Weatherly Foundry & Manufacturing Co., licensed by International Nickel Company.

Arsenic in Ternary Lead Base BEARING ALLOYS

By K. H. WEGNER*

Historical

ARSENIC is an element that is usually considered deleterious in most alloys. In 60:40 brass, for instance, as little as 0.12% reduces ductility as much as 50%[1]. In steel, amounts above 0.20% cause effects similar to that of phosphorus, namely, cold-shortness.[2] But in bearing metal alloys, it has been found to be a desirable element.

The literature dealing with arsenical bearing metals is not very comprehensive. Some investigations have been made by Roast and Pascoe[3] who show that a small amount of arsenic produces an extraordinary refining effect and has the tendency to transform the cubic crystals of antimony into needle-like plates. Also that arsenical alloys maintain their hardness better at higher temperatures. These conclusions were drawn from a study of alloys containing essentially lead, antimony and arsenic but no tin. Freeman and Brandt[4] show that the compressive strength and Brinell hardness is increased at normal and elevated temperatures in a lead base alloy containing 0.46% arsenic and in a tin base alloy containing as much as 3.00% arsenic. Most investigations have been made from the physical standpoint and it is the purpose of this paper to present a metallographical analysis of the effect of arsenic on the lead, antimony and tin system.

Method

An alloy similar to the A. S. T. M. grade No. 7 was selected for this investigation. This is a type used quite extensively in journal bearings and contains 75% lead, 10% tin and 15% antimony. A mixture of this composition was made of commercial metals and cast into a small bar. Pieces were cut from this bar and melted in a 100 cc. wrought iron crucible. No arsenic was added to the first melt. To the second melt, enough arsenic was added so that the fracture of the specimen revealed only a small amount of segregation; and finally, the third melt contained sufficient arsenic to eliminate practically all segregation. This series, designated as Group A (Fig. 1) was heated to about 700° F. and allowed to cool in the same crucible in which the melt was made. Brinell hardness tests, chemical analysis[5] and micrographs were made of the top and bottom portions of each specimen. These specimens were then remelted to 700° F. and poured into another crucible in which they were allowed to cool. This more rapidly cooled series, designated as Group B (Fig. 2) was tested in same manner as previous group. The results of all the tests are given in Table 1.

*C. M. St. P. & P. R. R. Co.

Discussion

Two important conclusions drawn from inspection of this table are, first, the alloys with insufficient arsenic lack homogeniety and must be composed of different structures as shown by the variation in Brinell hardness and chemical composition in same specimen and, secondly, the alloy with highest arsenic is practically uniform throughout.

Directing the attention now to macrographs of the

Table I

GROUP A

	A1		A2		A3	
	Top	Bottom	Top	Bottom	Top	Bottom
% Arsenic	0.13	0.09	1.68	0.31	1.69	1.54
% Antimony	18.57	11.47	15.28	12.21	15.13	15.06
% Tin	13.44	8.40	10.92	7.63	11.00	10.78
% Lead (Diff.)	67.86	80.04	72.12	79.85	72.18	72.62
Brinell No.	29.7	21.0	25.9	21.8	24.9	24.9

GROUP B

	B1		B2		B3	
	Top	Bottom	Top	Bottom	Top	Bottom
% Arsenic	0.13	0.08	1.58	1.15	1.79	1.44
% Antimony	17.40	10.65	15.72	16.45	15.35	15.20
% Tin	12.81	7.97	11.13	10.85	11.20	10.61
% Lead (Diff.)	69.66	81.30	71.57	71.55	71.66	72.72
Brinell No.	29.7	22.3	22.3	21.0	22.8	21.0

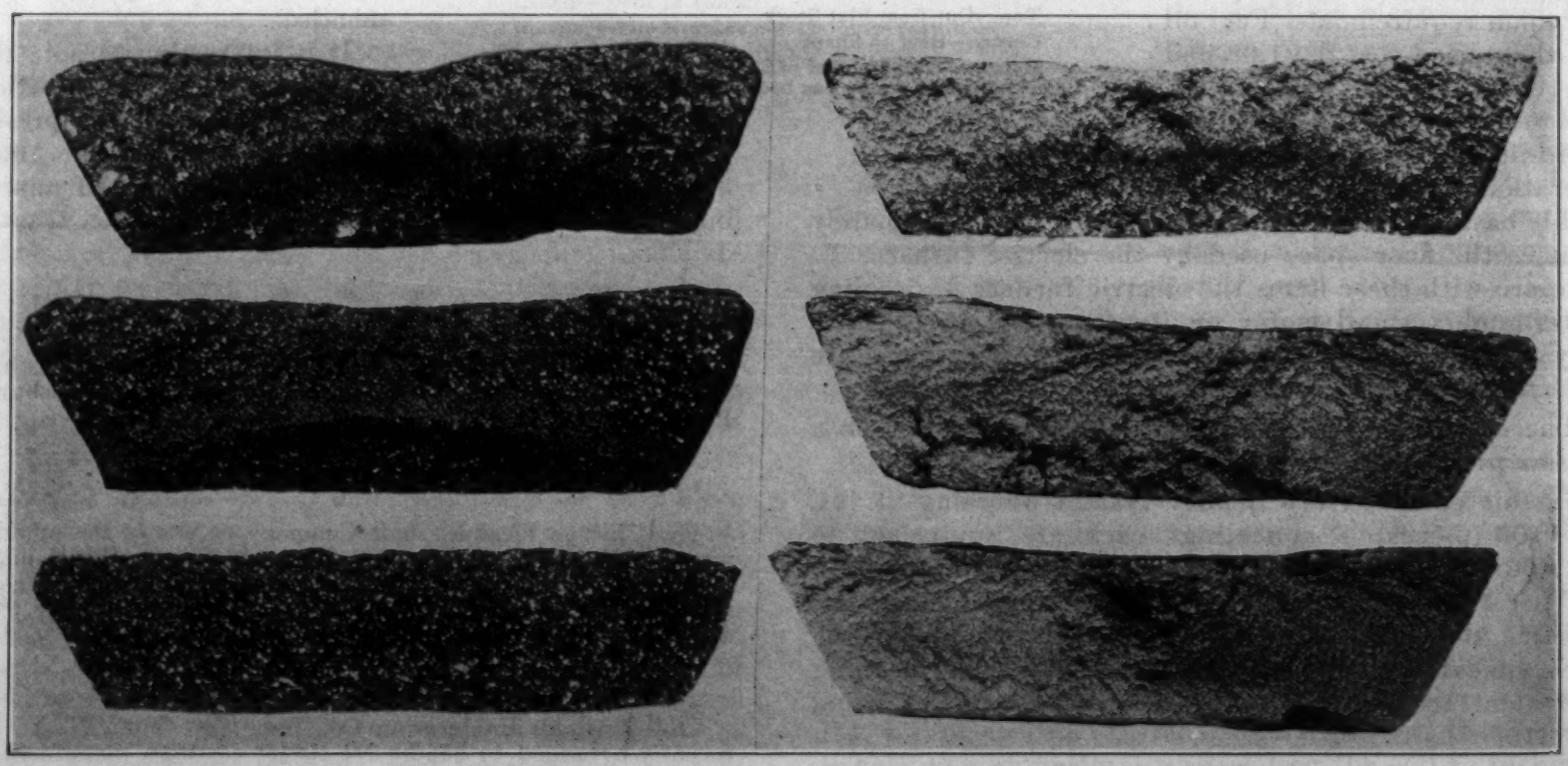


Fig. 1. ×1½. Macrograph Group A Specimens. Top—A1, Center—A2, Bottom—A3.

Fig. 2. ×1½. Macrograph Group B Specimens. Note grain refinement of arsenical specimens, B2 and B3. Top—B1, Center—B2, Bottom—B3.

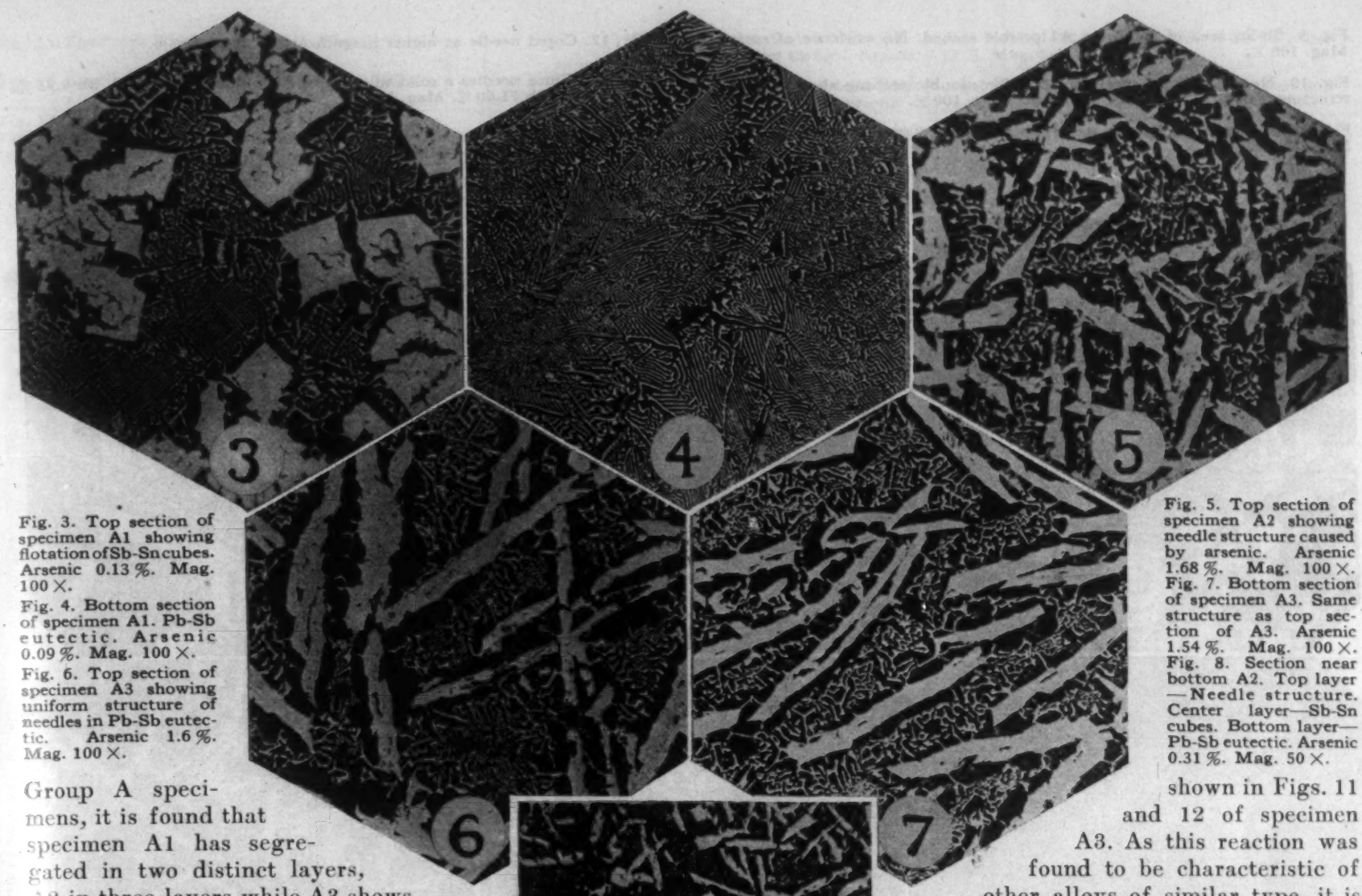


Fig. 3. Top section of specimen A1 showing flotation of Sb-Sn cubes. Arsenic 0.13 %. Mag. 100 ×.

Fig. 4. Bottom section of specimen A1. Pb-Sb eutectic. Arsenic 0.09 %. Mag. 100 ×.

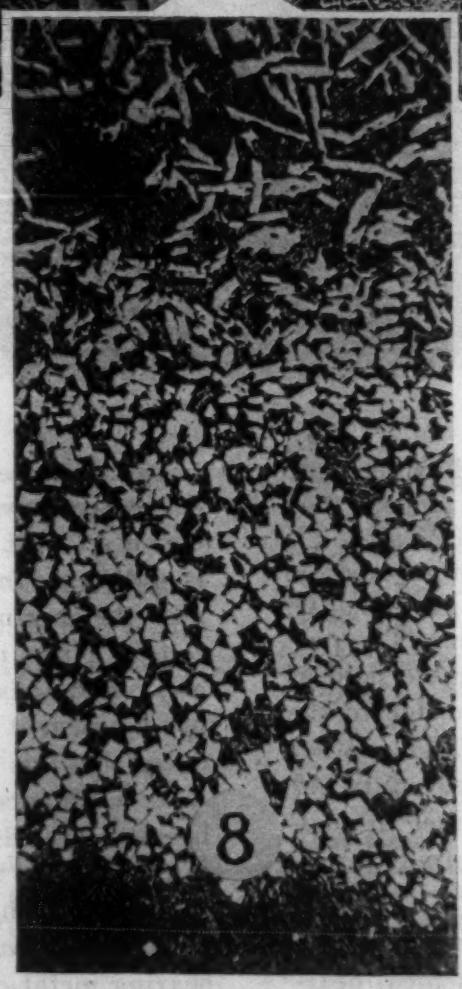
Fig. 5. Top section of specimen A2 showing needle structure caused by arsenic. Arsenic 1.68 %. Mag. 100 ×.

Fig. 6. Top section of specimen A3 showing uniform structure of needles in Pb-Sb eutectic. Arsenic 1.6 %. Mag. 100 ×.

Fig. 7. Bottom section of specimen A3. Same structure as top section of A3. Arsenic 1.54 %. Mag. 100 ×.

Fig. 8. Section near bottom A2. Top layer—Needle structure. Center layer—Sb-Sn cubes. Bottom layer—Pb-Sb eutectic. Arsenic 0.31 %. Mag. 50 ×.

Group A specimens, it is found that specimen A1 has segregated in two distinct layers, A2 in three layers while A3 shows a uniform structure. Further examination by aid of the microscope reveals the nature of these layers. Fig. 3* is a micrograph of specimen A1 which shows that the upper layer is due to the flotation of Sb-Sn crystals. These being hard and brittle cause a higher Brinell number than the softer Pb-Sb eutectic which constitutes the lower layer, Fig. 4. Considering now specimen A2, it is found that in addition to eutectic and Sb-Sn crystals which form the lower and center layers, respectively, there is a new type of structure in the upper layer, Figs. 5 and 8. Apparently the arsenic has in some manner, caused a number of the Sb-Sn crystals to change from cuboidal to needle-shaped plates. When the arsenic is present in sufficient quantity to affect practically all of the Sb-Sn crystals, the uniform structure shown by Figs. 6 and 7 is obtained. Obviously, it is the interlocking effect of these needles that prevents segregation and produces homogeneity. The fact that arsenic seemed to be associated in some manner with these needles, led to the conclusion that if such specimens were etched with a suitable reagent it would be possible to detect the arsenic-rich areas microscopically. Consequently, several of these specimens previously etched with Vilella-Beregekoff reagent, were immersed 15 to 30 seconds in 1:1 ammonium poly-sulphide solution. The effect of this double etching on the Sb-Sn crystals is shown by Fig. 9. Except for some staining, there appeared to be no other change. The needle crystals, however, were found to have a centralized core, Fig. 10. This structure is more clearly shown in Figs. 11 and 12 of specimen A3. As this reaction was found to be characteristic of other alloys of similar type, it is presumed that these central areas are rich in arsenic and are made visible by the selective action of the ammonium poly-sulphide. The peculiar location of these areas may be explained by Mansuri's investigations of the arsenic-tin[6] and arsenic-antimony[7] systems. He has found that arsenic and tin form an α solid solution up to 29.39% arsenic; and that arsenic and antimony form a solid solution in all proportions. Fig. 13 shows some white needles of α solid solution in an alloy containing 3.05% arsenic, 0.95% antimony, 24.32% tin and 71.60% lead. Fig. 14 is an example of the effect of arsenic on the antimony cubes in an alloy containing 5.58% arsenic, 22.26% antimony and 72.13% lead. It will be noted that in both cases the solid solutions assume the form of needles. These specimens did not show any evidence of coring when double etched. This would indicate that both tin and antimony must be present with arsenic to produce this type of structure. Evidently, then, the cores are composed of all 3 of these elements probably as a ternary solid solution. It is believed that this solid solution forms primary needle-shaped crystals around which the Sb-Sn compound precipitates resulting in the formation of cored needles instead of the usual cubes.



In rather quickly cooled arsenical babbitts, it is evidently this process of crystallization that causes the formation of an infinite number of nuclei throughout the freezing mass thus producing the fine structure shown by Figs. 16 and 17 and macrograph B. As will be remembered this series was produced by pouring the molten

*All samples etched with Vilella-Beregekoff Reagent:
1 part Nitric Acid (Conc.)
1 part Acetic Acid (Conc.)
4 parts Glycerine.

Fig. 9. Sb-Sn area of specimen A2 double etched. No evidence of coring. Mag. 100 ×.

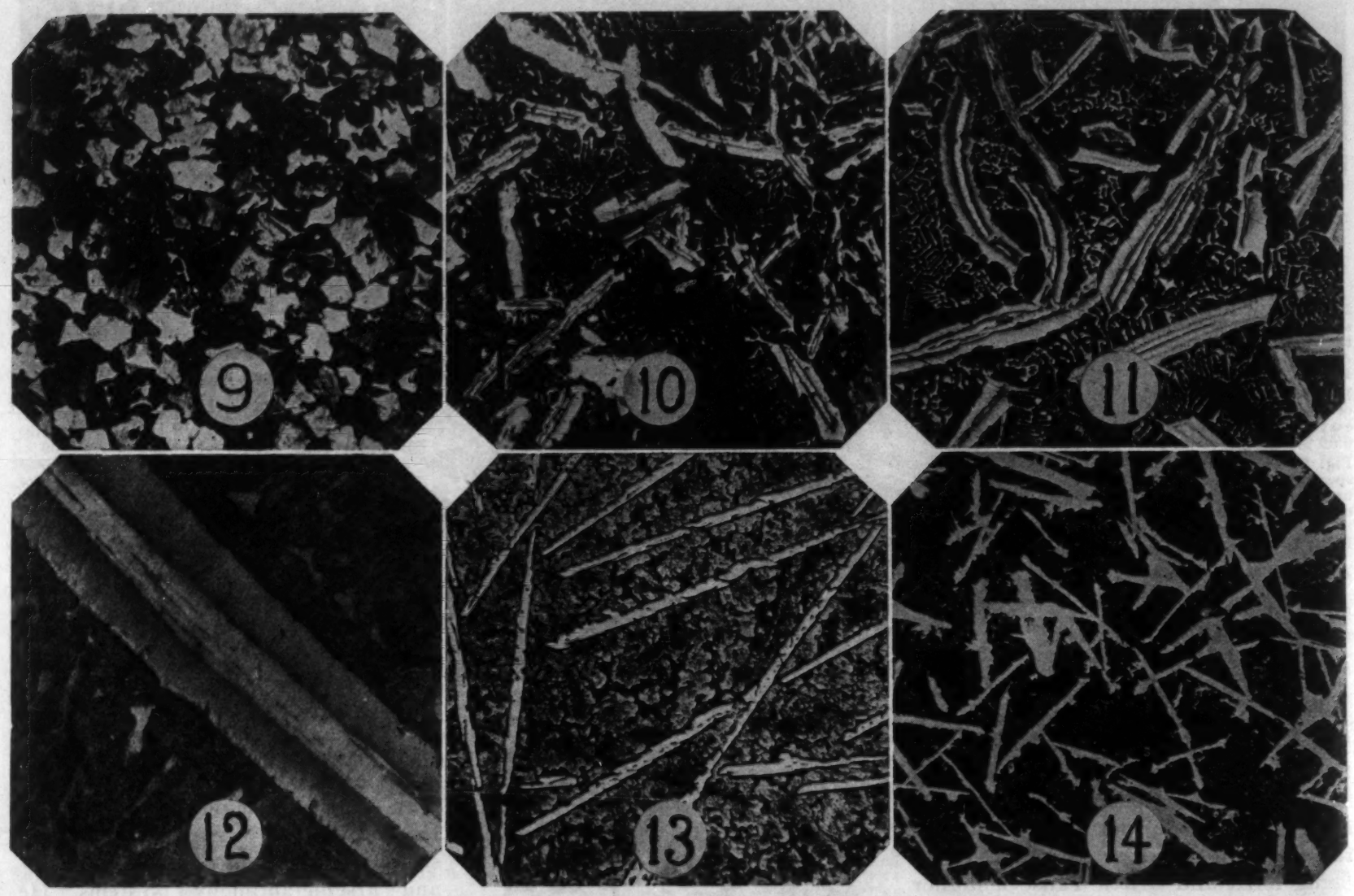
Fig. 10. Needle area of specimen A2 after double etching showing cored structure, due to the presence of arsenic. Mag. 100 ×.

Fig. 11. Specimen A3 double etched showing cored structure more clearly. Mag. 100 ×.

Fig. 12. Cored needle at higher magnification. Mag. 500 ×.

Fig. 13. White needles a solid solution of As and Sn. As 3.05 %, Sb 0.95 %, Sn 24.32 %, Pb 71.60 %. Mag. 100 ×.

Fig. 14. White needles—As-Sb solid solution. As 5.58 %, Sb 22.26 %, Pb 72.13%. Mag. 100 ×



metal from one crucible to another and allowing it to cool therein. This procedure did not have much effect on the structure of specimen A1 other than causing the formation of somewhat smaller cubes and more uniform distribution of the constituents in the upper portion, Fig. 15, while the lower portion remained the same. The persistent formation of blow holes in the segregated field of both specimens A1 and B1 is believed to be due to the contraction of the Pb-Sb eutectic on cooling. The rather uniform structure of specimen B2 as compared to A2 (see Table 1) is obviously due to the difference in cooling rate which in this case was sufficiently rapid to prevent flotation of the arsenic free Sb-Sn cubes.

This discussion would not be complete unless some mention were given of the advantages of arsenical bearing metals. The most important is that segregation is eliminated which, if present, not only causes considerable trouble in chemical determinations, but may also contribute to babbitt failure.

It is a known fact that a good bearing metal should have hard particles uniformly distributed in a soft matrix. Where segregation has occurred, this is not the case and undoubtedly the unequal distribution of the hard particles causes localized overheating and consequent failure. This condition is illustrated by Figs. 18 and 19 which are examples of lead base babbitts that have failed in service. Even in chill cast lead-base alloys segregation occurs as shown by Fig. 20 which is a micrograph of specimen A1 cast in a steel mold. The lack of homogeniety is quite evident. Although this condition may not occur near the bearing surface of new bearings, failure is likely to take place, nevertheless, as the lining gradually wears away and these areas are reached. Even if this should not happen, it is probable that the finer structures produced by this rapid cooling would result in decreased wear resistance. It has been shown that in the case of bearing bronzes, the fine structures wear out more rapidly than coarse material.[8] It would not be unreasonable to suppose that this would be true of babbitt structures also. By adding arsenic to the lead-base alloys, coarse uniform bearing metal can be obtained without the attendant evil of segregation or decreased wear resistance.

Conclusion

This investigation has shown that arsenic has a pronounced effect on the crystalline structure of lead-base alloys. It forms needle-shaped crystals containing an arsenic-rich core consisting, probably, of a ternary solid solution of arsenic, antimony and tin. The interlocking of these needles prevents segregation and permits the production of uniform alloys of the lead-base type under variable conditions. The determination of cooling curves,

Fig. 15. Specimen B1. Arsenic 0.13 %. Mag. 100 ×.

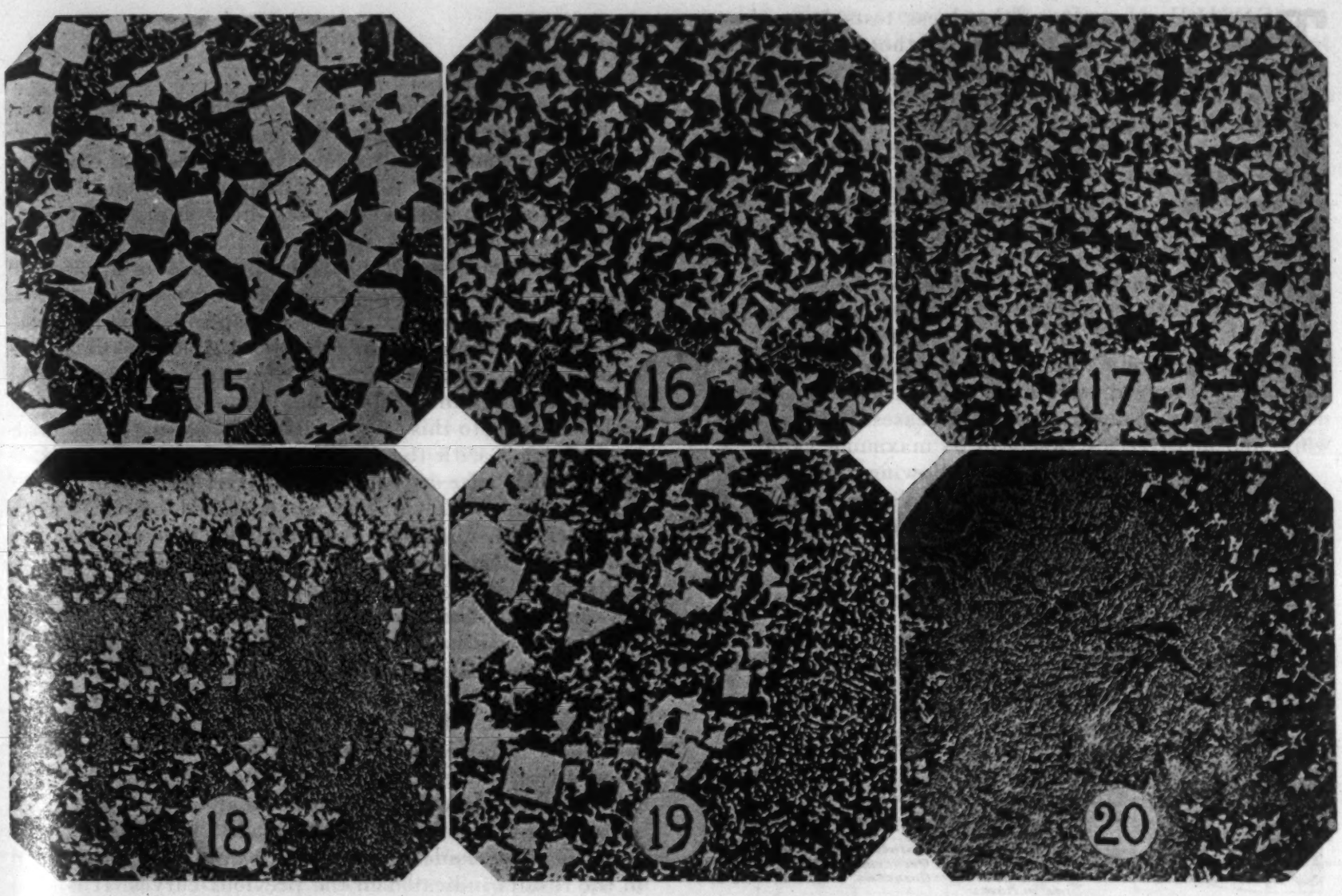
Fig. 16. Specimen B2. Arsenic 1.58 %. Mag. 100 ×.

Fig. 17. Specimen B3, Arsenic 1.79 %. Mag. 100 ×.
Photomicrographs of group B series showing effect of small change in cooling rate on crystalline structure of specimens A1, A2, A3.

Fig. 18. Failed lead base journal Babbitt. Note thick skin of Sb-Sn cubes at bearing surface. Arsenic 0.13 %. Mag. 50 ×

Fig. 19. Failed lead base journal Babbitt. A typical example of flotation occurring in low arsenic alloys of this type. Arsenic 0.20 %. Mag. 100 ×.

Fig. 20. Specimen A1 chill cast showing unequal distribution of Sb-Sn cubes. Arsenic 0.13 %. Mag. 50 ×.



wear resistance, physical properties and a study of the As-Sb-Sn system would aid materially to a more complete understanding of arsenical babbitts.

[1] C. J. Smithells. Impurities in Metals. John Wiley & Sons, New York, 1928.
[2] The Making, Shaping & Treating of Steel. Handbook, Carnegie Steel Co. 4th Edition, 1925.
[3] H. J. Roast & C. F. Pascoe. How Arsenic Affects Alloys. *Foundry*, Vol. 51, Jan. 15, 1923, page 67.
[4] Freeman & Brandt. The Effect of Impurities on Compressive Strength and Brinell Hardness of Babbitt. *Proceedings American Society for Testing Materials*, Vol. 24, Part 1, page 253.
[5] Price & Meade. Technical Analysis of Brass and the Non-ferrous Metals. John Wiley & Sons, New York, 1917. 2nd Edition.
[6] Q. A. Mansuri. Arsenic-Tin System. *Journal Chemical Society*, Vol. 123, Part 1, page 214.
[7] Q. A. Mansuri. Arsenic-Antimony System. *Journal Chemical Society*, Vol. 131, Part 1, page 2107.
[8] H. J. French. Wear and Mechanical Properties of Railroad Bearing Bronzes at different Temperatures. *Bureau of Standards, Journal of Research*, Vol. 1, 1928, page 343.

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The Electric Furnace Company of Salem, Ohio, has purchased the business of the Process Engineering & Equipment Corp., Attleboro, Mass., builders of the "Preeco" bright annealing and controlled atmosphere furnaces. All equipment, etc., of the Process Corp., has been moved to Salem, Ohio.

The Process Engineering & Equipment Corp., has been specializing on bright annealing and controlled atmosphere furnaces employing a principle which is claimed to prevent oxidation and the formation of scale by the use of a gas, a mixture of gases, or a vapor to exclude air during heating and cooling. The material is thus removed from the furnace in a bright, clean and dry condition ready for succeeding operation of finishing, thus eliminating dipping, pickling, cleaning, drying, etc.

The acquisition of the Process Engineering & Equipment Corp., together with the Electric Furnace Company's own newly developed bright annealing process, enables them to offer both large and small production units with controlled atmosphere for tube, sheet, strip, coil, wire, etc., both ferrous and non-ferrous.

Atlas Electric Devices Co., Chicago, manufacturers of apparatus for predetermining the resistance of materials to sunfading, weather and washing, announces the appointment as its Pacific Coast Representatives of L. H. Butcher & Co., 274 Brannan St., San Francisco, with branches in Los Angeles, Portland and Seattle.

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The Stibloy Products Company, Inc., with principal offices in the Koppers Building, Pittsburgh, has taken over the assets of Liquid Metal Products, Inc., Chicago, producers and distributors, under the Arent patents, of Stibloy, a metal compound in liquid form, which acts as a primary coating to hold paint, enamel and lacquer tenaciously, permitting immediate finishing of new galvanized metal surfaces.

Stibloy is claimed to extend the life of galvanized surfaces by protecting them from the effects of atmospheric conditions and from the damage caused by exposure to gases, acid fumes, smoke and brine. It is used for protecting and preserving galvanized roofing, siding, sheeting, guttering, downspouts, wire fencing, air ducts, car roofing, screens, nails, transmission towers and other galvanized products . . . as a primer, not a paint.

Hardness Change of Duralumin

By FRANZ

TENSILE strength and hardness tests have shown themselves of value in following the change in properties of Duralumin during self-aging after having been quenched from temperatures of between 500° and 535° C. The hardness determinations in such examinations have heretofore been made almost exclusively with the Brinell hardness tester.

This change in hardness with time as measured with the Brinell hardness tester is shown in a typical curve of Fig. 1. The hardness tests in this case were made on samples 2 mm. thick of rolled samples of the alloy 681 B using a 2.5 mm. steel ball and a load of 62.5 kg. The samples were first heated for 30 minutes at 500°, 515° and 530° C. respectively and then quenched in water at room temperature. Immediately after the quenching the hardness decreases somewhat, then increases rapidly for a while and finally rises slowly to its maximum value.

Fig. 2 shows the results of following this hardness change with the Rockwell Hardness Tester using a 1/8" steel ball and a 10 kg. initial load and a 100 kg. test load. It is seen that this test shows essentially the same kind of hardness change as is indicated by the use of the Brinell tester.

*Aachen, Germany.

†Translated by Raymond H. Hobrock, Chicago, Illinois. Original published in *Abhandlungen aus dem Aerodynamischen Institut an der Technischen Hochschule zu Aachen*, No. 10, 1931, Verlag Julius Springer, Berlin.

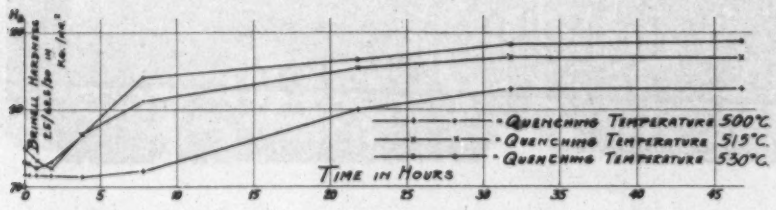

Fig. 1. Influence of Quenching Temperature on Change of Brinell Hardness During Aging of Duralumin at Room Temperature.

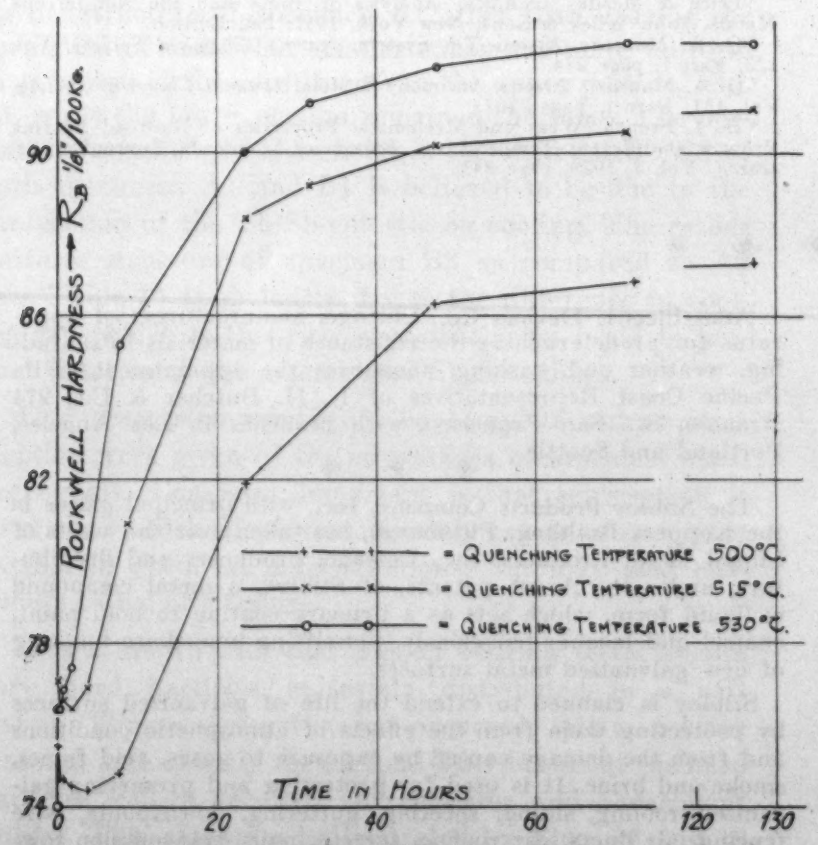

Fig. 2. Influence of Quenching Temperature on Change of Rockwell Hardness During Aging of Duralumin at Room Temperature.

This correspondence is to be expected since the physical conditions under which the measurements were made are approximately the same in both cases; the diameter of the balls used are not much different from one another and the loads applied are to each other approximately as the square of the diameters of the balls. Therefore about the same specific surface pressure is applied in each case. For example, using the above indicated ball diameter and pressures, where the specific surface pressure of the Brinell tester is 105 kg./mm.^{2} the corresponding specific surface pressure in the case of the Rockwell tester in 104 kg./mm.^{2}

In contrast to this the results of following these hardness changes with the Herbert pendulum are quite different. For these measurements a Herbert pendulum of 4 kg. weight supplied with an unetched ball of 1 mm. diameter was used. The measurements in every case are indicated in terms of the Herbert time hardness and is the time required for 10 simple swings of the above described pendulum in seconds of time. Each indicated observation is the average of 6 separate determinations. Before these determinations were made the pendulum was so adjusted as to indicate a time hardness Zs—100 seconds on the glass plate supplied with the pendulum for purposes of calibration. Fig. 3 shows the results of these time-hardness measurements on samples that had been given exactly the same heat treatment as those from which Figs. 1 and 2 were obtained. The measurements actually were made on the same samples as those used in the results indicated in the previous curves. It is seen that the results are considerably different than those obtained by measurement with the Brinell or the Rockwell tester. As in the previous measurements, the hardness decreases immediately after the quench followed by a rapid increase, this time, however, to a value which is the highest measured at any time. Thereafter the hardness values decrease and approach a constant value with time.

These results are the more remarkable in the light of the fact that there is usually a definite correspondence between the hardness values as determined by the Brinell, Rockwell and Herbert testers on pure metals

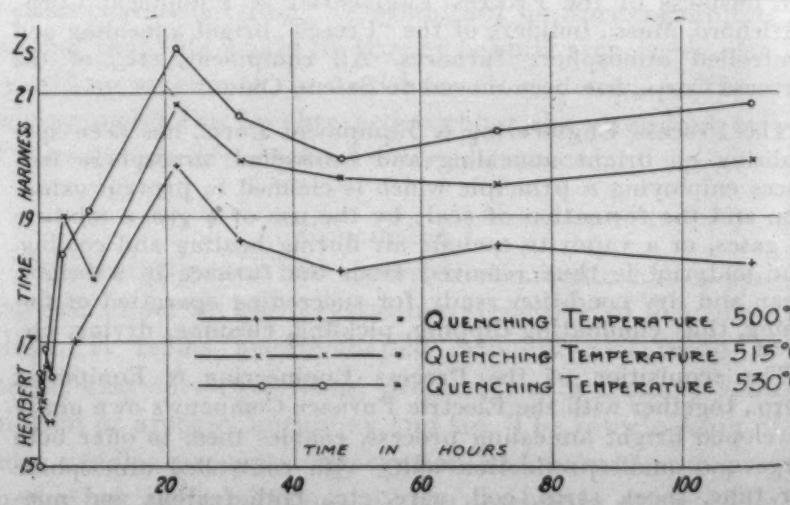

Fig. 3. Influence of Quenching Temperature on Change of Herbert-time-hardness During Aging of Duralumin at Room Temperature.

with Time and at Different Depths†

BOLLENRATH*

and on technical ferrous and non-ferrous alloys.[1] However the shape of the curves are very much like those obtained by the author[2] by following the aging of Lautal at 1600° C. with the Brinell tester and the aging of an Al-Cu-Ni alloy with the Rockwell tester as is shown in Figs. 4 and 5.

When the Herbert pendulum is used, the measured hardness is more influenced by the elastic deformation than in the case of the Brinell tester, although in the measurement of hardnesses such as are here involved this portion due to elastic deformation is small.[3] The differences in the results of the 2 hardness measurement methods is, therefore, not explainable on these grounds. There is however a considerable difference in the 2 methods of hardness testing and that exists in the fact that due to the very small ball and the small load used in the Herbert pendulum the volume of material affected in making one measurement is considerably less than in the case of either the Rockwell or the Brinell tests. The depth effect of the Herbert test, therefore, is considerably less than with the other tests. However, the specific surface pressure in the case of all the testing methods is of the same order of magnitude. For example, in the case where the specific surface pressure in the Brinell test is 105 kg./mm.^{2} the corresponding force for the Herbert tester is 76 kg./mm.^{2} This difference in the depth effect between the Brinell and the Herbert pendulum hardness tests has recently found application in the determination of the thickness of the case of nitrided samples without sectioning them.[4] All of these considerations lead one to suspect that there is no uniform hardness throughout the thickness of the Duralumin samples previously described even though these samples are very thin. In order to investigate this, samples were prepared according to Fig. 6 which permitted the measurement of the Herbert pendulum hardnesses throughout the cross section. Fig. 7 shows the result of measurement of a sample of Duralumin as obtained from the factory (without any sort of additional heat treatment) and shows that there actually are hardness differences at depths that would be entirely indistinguishable in using the Brinell tester because of its much greater depth effect. The Brinell test, therefore, yields only an average result such as is also obtained in a tensile strength test or in a resistance measurement test. If the samples investigated should be made thick enough it might be expected that even the Brinell test would show these hardness differences with depth. Step bars shown in Fig. 8 were, therefore, prepared of pressed Duralumin bars 30 × 30 mm. in section. In making the steps in the quenched samples great care was exercised to remove only extremely thin layers of the metals at one time in order to avoid heating the metal. As an additional precaution the last portion of the metal on each step was removed by fine polishing paper in order

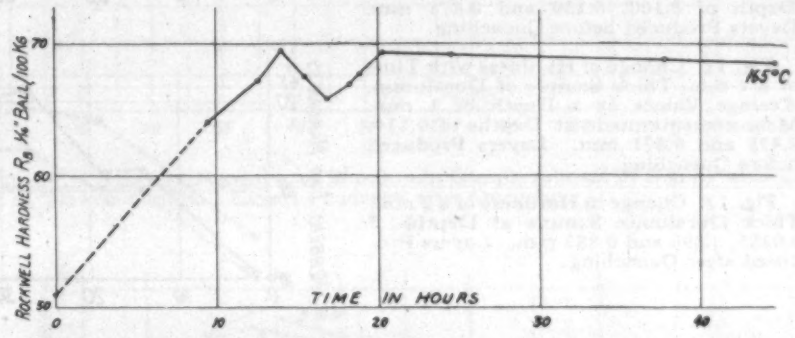

Fig. 5. Change in Rockwell Hardness of U 11 During Aging at 165°C Quenching Temperature 535° C.

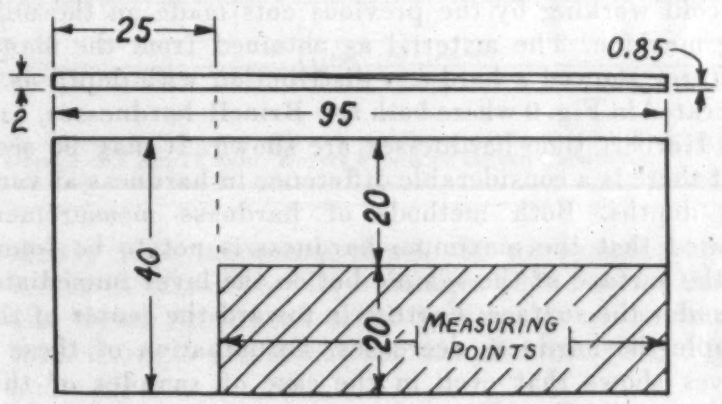

Fig. 6. Samples Used for Experiments. Results shown in Fig. 7.

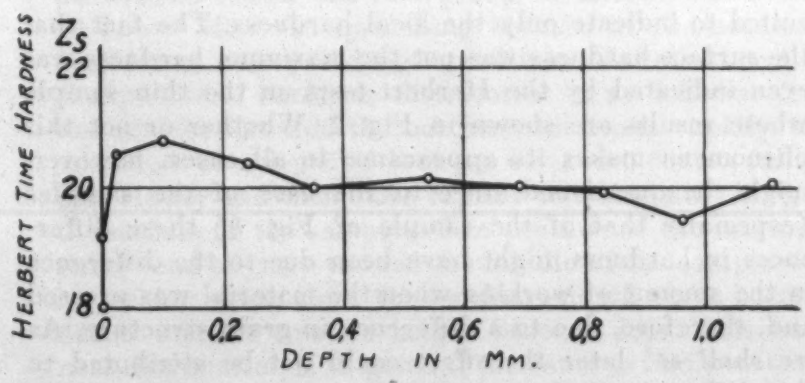

Fig. 7. Change of Hardness with Depth on 2 mm. Thick Sample shown in Fig. 6. Sample as Received, No Heat Treatment.

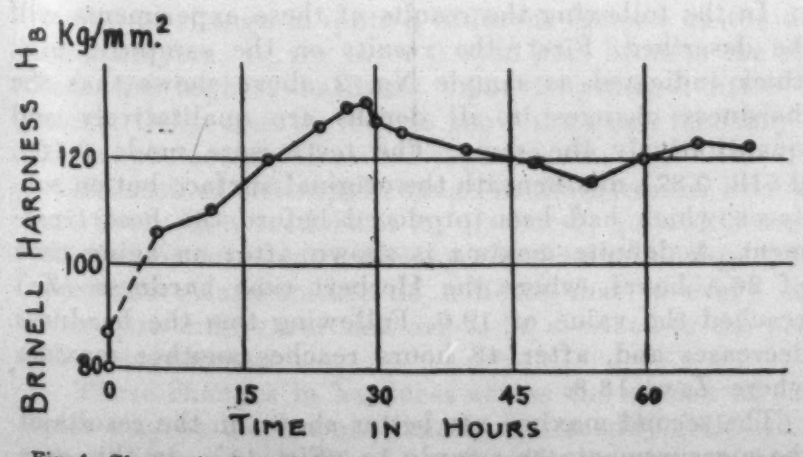

Fig. 4. Change in Brinell Hardness of Lautal During Aging at 160°C. Quenching Temperature 510° C.

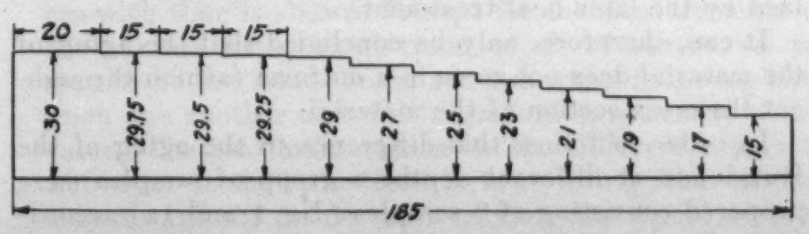

Fig. 8. Dimensions of Samples Used for Experiments in mm. Results shown in Fig. 9.

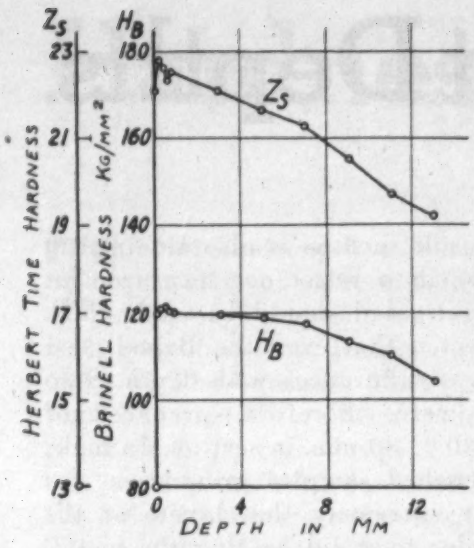

Fig. 9. Distribution of Hardness with Depth in the Initial Condition. (Samples as shown in Fig. 8.)

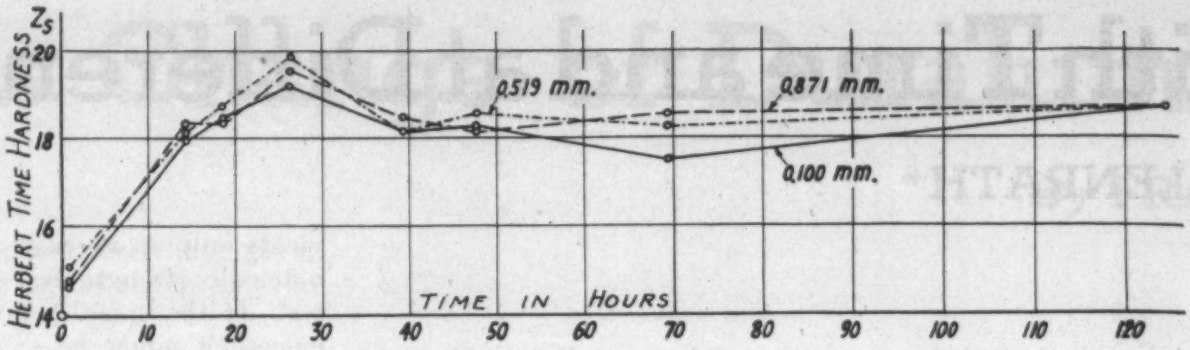

Fig. 10. Change in Hardness with Time of a 2 mm. Thick Duralumin Sample at Depth of 0.100, 0.159 and 0.871 mm. Layers Produced before Quenching.

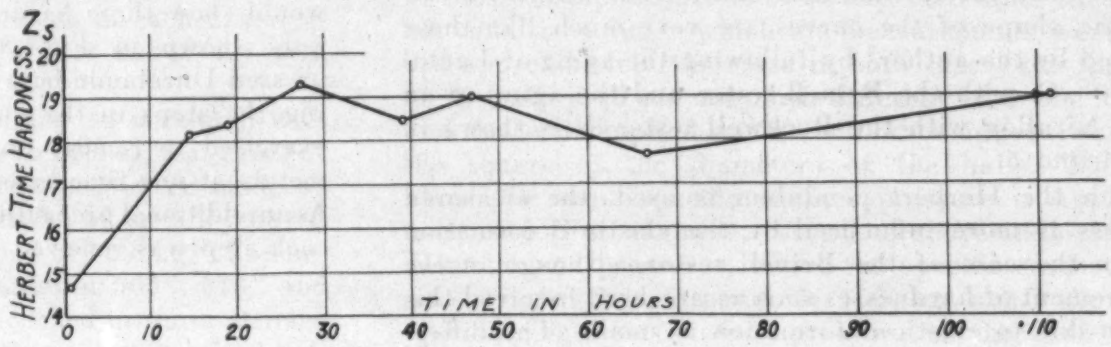

Fig. 11. Change of Hardness with Time of a 2 mm. Thick Sample of Duralumin, Average Values at a Depth of 1 mm. Measurements made at Depths of 0.114, 0.475 and 0.871 mm. Layers Produced before Quenching.

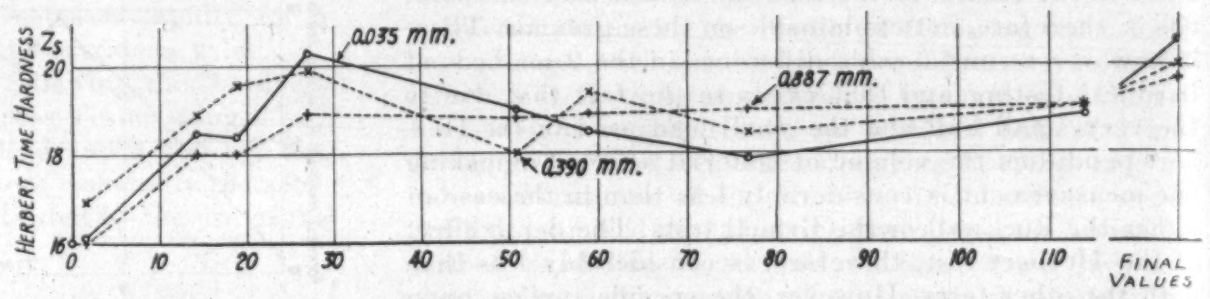

Fig. 12. Change in Hardness of a 2 mm. Thick Duralumin Sample at Depths of 0.0355, 0.390 and 0.887 mm. Layers Produced after Quenching.

to remove in this way any possible hardening effect due to cold working by the previous cuts made on the milling machine. The material as obtained from the manufacturer showed a hardness distribution with depth as is indicated in Fig. 9 where both the Brinell hardnesses, and the Herbert time hardnesses are shown. It may be seen that there is a considerable difference in hardness at various depths. Both methods of hardness measurement showed that the maximum hardness is not to be found on the surface of the sample but on the layer immediately under the surface. Farther in toward the center of the sample the hardness decreases. Examination of these 2 curves shows that even in the case of samples of this thickness the Brinell test gives an average result of the hardnesses extending over a considerable thickness of the metal. Here too, the Herbert tester is much more suited to indicate only the local hardness. The fact that the surface hardness was not the maximum hardness was even indicated by the Herbert tests on the thin sample whose results are shown in Fig. 7. Whether or not this phenomena makes its appearance in all cases, however, might be questioned since in the case of the samples (especially that of the sample of Fig. 8) these differences in hardness might have been due to the difference in the amount of working when the material was pressed and, therefore, due to a difference in grain structure. As we shall see later the effect could not be attributed to this difference in the grain structure, especially since in other samples the grain structure was largely homogenized by the later heat treatment.

It can, therefore, only be concluded that the aging of the material does not go on in a uniform fashion throughout the cross section of the material.

In order to follow this difference in the aging of the Duralumin at different depths a group of samples were prepared consisting of 2 samples (No. 1 and 1a) according to Fig. 6, another unmachined sample from the same sheet of metal as 1 and 1a (called No. 2), a machined sample according to Fig. 8 (No. 3) and an unmachined piece of material of the 30 × 30 mm. bar (No. 4). These samples were all heated to 515° C., maintained there for 30 minutes and then quenched in water at room temperature. Immediately after this heat treatment the samples (No. 2 and No. 4) which had not been machined to shape were given the shapes indicated in Figs. 6 and 8 respectively and the surfaces of all of the samples were tested with the Herbert pendulum as previously described. Thus by comparing the results of the samples machined before and those machined after the heat treatment it would be possible to detect any differences in hardness that might result from differences in structure at the different depths. It was considered of importance to accurately determine the increase in hardness that follows immediately after the initial decrease in hardness and after about ½ to ¾ hour aging,[2] the formation of the maximum hardness which is to be expected about 15 to 25 hours after the beginning of the increase in hardness,[3] and the course of the hardness changes immediately after the maximum. Therefore, in order to obtain accurate information at these points many more measurements were made.

In the following the results of these experiments will be described. First: the results on the sample 2 mm. thick indicated as sample No. 2 above shows that the hardness changes in all depths are qualitatively and quantitatively the same. The tests were made 0.100, 0.519, 0.871 mm beneath the original surface but on surfaces which had been produced before the heat treatment. A definite maxima is shown after an aging time of 26.5 hours, where the Herbert time hardness (Zs) reached the value of 19.6. Following this the hardness decreases and, after 48 hours reaches another maxima where Zs = 18.3.

The second maxima are better shown in the results of the measurements on sample 1a (Fig. 11)—in this case it appears after an aging time of 47 hours. After this

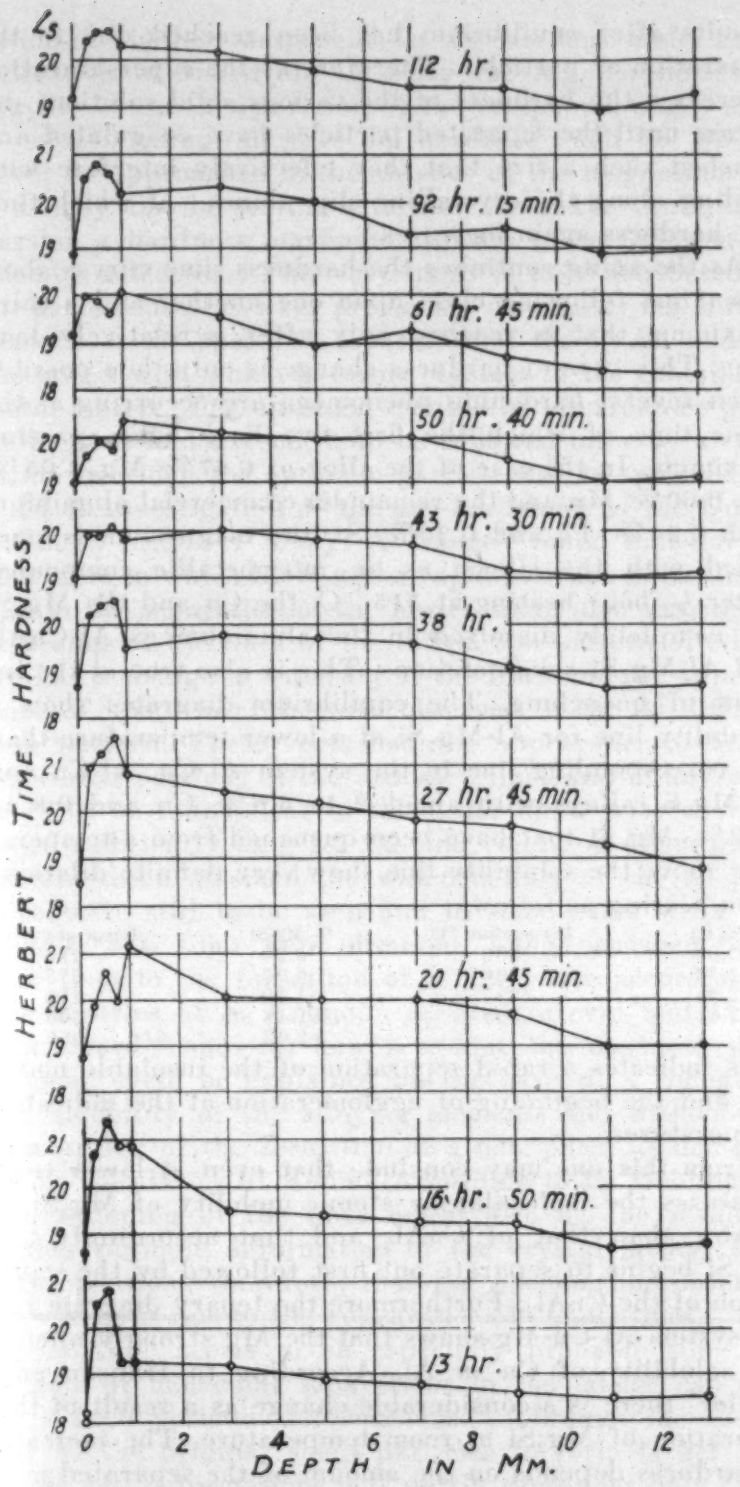

Fig. 13. Distribution of Hardness Across the Section of Sample Shown in Fig. 8 after Various Periods of Aging. Layers Produced After Quenching.

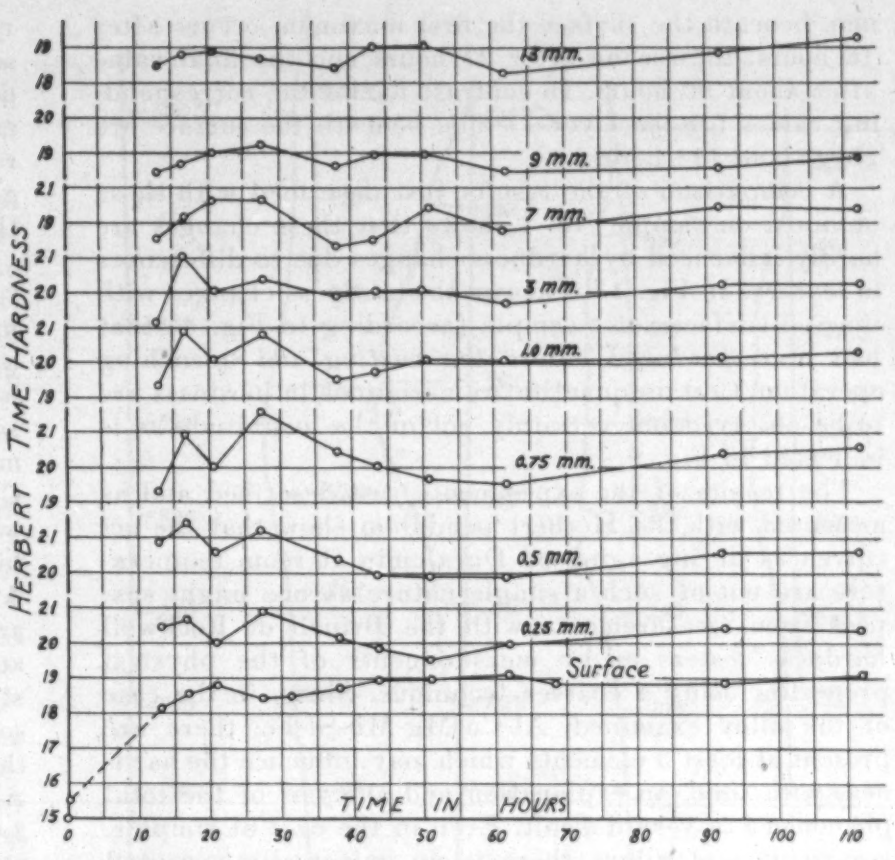

Fig. 14. Change in Hardness of Duralumin Sample (shown in Fig. 8) with Time and at Various Depths. Layers Produced after Quenching.

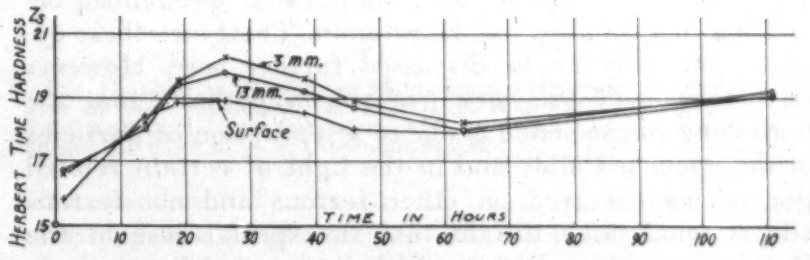

Fig. 15. Change in Hardness of Duralumin Sample (Fig. 8) With Time on Surface and at Depths of 3 and 13 mm. Layers Produced Before Quenching.

the hardness falls again to a value of Zs = 17.6, and at 65 hours increases slowly toward the final value which is still not attained after 112 hours.

The course of this hardness change of the sample No. 2 with time is shown in Fig. 12 where these hardness changes on the different depth layers (produced after the heat treatment) are indicated. The hardness is observed to change in quite a different fashion on the different layers. In the layer 0.0355 mm. beneath the surface the highest hardness value is attained after 27 hours. Otherwise the curves show the same tendency as in the previously described experiments.

Because of the rapid change in the hardness over the relatively thin sections it is, of course, quite difficult to obtain measurements showing the gradual change, however the measurements do indicate that in every case there are differences that are to be ascribed to the place of measurement.

These changes in hardness across the section of samples may be much more exactly and readily followed on the thicker samples Nos. 3 and 4. The layers are thicker and, therefore, the local hardness changes much more slowly with increasing depth. If we examine how the hardness of sample No. 4 changes with time at the various layers (produced after quenching) we see that the total changes are somewhat complicated. (See Figs. 13 and 14.)

In Fig. 13 the hardness distribution over the cross section for various times of aging are shown. It was not possible to obtain the first group of measurements until 13 hours after the quenching because of the necessary machining and polishing operations. At all times the characteristic hardness distribution is as follows: from the surface inwards the hardness increases rapidly to a maximum at a depth of between 0.4 and 0.6 mm., falls at first slowly and then more rapidly to a minimum at about the middle of the samples. The position of the maximum value tend slightly toward the middle of the sample with increasing aging time, the hardness-depth curve becomes flatter and its relative height becomes less. This hardness distribution is in general the same as had previously been found (Fig. 9).

In Fig. 14 the change in hardness of the various layers with time is shown. Except for the measurements on the surface the general trend of all of the curves are seen to be the same. There are 2 maxima following close upon one another and then a gradual approach to a final value. The maximum values on the different layers do not occur at the same time. It is seen that from the outside toward the inside the 2 maxima are separated more and more from one another and apparently need longer times for their formation. For example on the layer 0.25

mm. beneath the surface the first maximum occurs after 16 hours, the second after 27 hours and the final value after about 90 hours. In contrast to this the corresponding values for the layer 11 mm. beneath the surface are 28, 51 and 114 hours.

A comparison of the results just described with those obtained on sample No. 3 shows that these changes are hardly influenced by hardness changes due to differences in texture. In Fig. 15 is shown the hardness changes with time on surfaces of a sample (according to Fig. 8) that had been produced before the heating and quenching operation. Certain quantitative variations in hardness are to be observed but certainly not of the magnitude as is indicated in Fig. 9.

The results of the experiments just described and as measured with the Herbert pendulum show that the occurrences during aging of Duralumin at room temperature are not of such a simple nature as one might suspect from measurements with the Brinell or Rockwell hardness testers or by measurements of the physical properties using a coarser technique. Since, in the case of the alloy examined, Al-Cu-Mg-Mn-Si-Fe, there are present at least 3 elements which may influence the hardness with time, an explanation and analysis of the total phenomena is very difficult. Even in the case of simpler, two-component alloys there is no universally accepted explanation of the machinism of the change in hardness with time. Much less, therefore, is there an accepted explanation of the much more complicated mechanism occurring in such alloys as Duralumin. Therefore these explanations will not be discussed further here. However on the basis of the very probable hypothesis that the hardening phenomenon is due to a separation of particles in the quenched alloy and in the light of certain related phenomena observed on other ferrous and non-ferrous alloys, some small insight into the special case of the Duralumin age-hardening will be attempted.

In regard to the time-hardness curves there arises the question as to what causes the decrease in hardness immediately after the quenching. Undoubtedly cooling stresses are present since it is impossible to cool all parts of the sample with the same speed when it is quenched. Furthermore one can assume that these cooling stresses are unequally distributed according to whether or not the grains are so oriented as to permit a more or less rapid conduction of heat.

In this way a deformation may result which, for the same reasons, is only local or at least varies with the orientation of the grains. These kinds of deformations, however, do not increase the lattice energy. It is rather to be expected that the conditions of the forces which lead to slippage, or to permanent deformations, are more readily attained with the assistance of these local forces due to cooling strains than without them.

In the case of Duralumin, as in the case of other similar alloys which age-harden at room temperature, a separation of the portions of those components of the alloy which are present in an unstable supersaturated condition commences immediately after the quenching. This initial separation might well relieve to a considerable extent the local stresses and this, in conjunction with the newly formed, highly disperse particles, might readily affect the electrical conductivity of the metal; on the other hand, it is also quite possible that these local stresses increase the speed and the amount of the separating particles as Dean, Day and Gregg have shown to be the case in the system Iron-Nitrogen.[5] There remains, therefore, but one conclusion to be drawn and that is that the strongly supersaturated solid solutions existing immediately after the quenching have a greater hardness than results after equilibrium has been reached due to the separation of particles. Therefore as the supersaturation decreases the hardness of the various solid solutions decrease until the separated particles have coagulated and reached such a size that they effectively interfere with gliding along the crystalline slip planes—at which time the hardness again increases.

As the aging continues the hardness time curves show 2 maxima following close upon one another and a third maximum that is reached only after a relatively long time. This type of hardness change is only then possible when several hardening phenomena are occurring at the same time of which the first two have but a passing maximum. In the case of the alloy of 0.47% Mg, 4.05% Cu, 0.60% Mn and the remainder commercial aluminium with 0.24% Fe and 0.30% Si, the magnesium is combined with the silicon as an intermetallic compound. After ½ hour heating at 515° C. the Cu and the Mg_2Si are completely dissolved in the aluminum as Al-$CuAl_2$ and Al-Mg_2Si solid solutions. This is also true at the instant of quenching. The equilibrium diagrams show a solubility line for Al-Mg_2Si at a lower temperature than the corresponding line in the system Al-Cu, Al-Cu and Al-Mg_2Si alloys with about 2 to 3.6% Cu and 0.8 to 2.02% Mg_2Si that have been quenched from a temperature above the solubility line show very definite dilations upon heating as follows:

% Cu	Temperature °C.	% Mg_2Si	Temperature °C.
2.05	350	0.80	280
2.34	330	0.93	275
2.95	315	1.10	270
3.62	300	1.48	260
		2.02	250

This indicates a rapid separation of the insoluble material and the beginning of agglomeration at the indicated temperatures.

From this one may conclude that even at lower temperatures the molecular or atomic mobility of Mg_2Si is greater than that of $CuAl_2$ and that accordingly the Mg_2Si begins to separate out first followed by the separation of the $CuAl_2$. Furthermore the tenary diagram for the system Al-Cu-Mg shows that the Mg strongly affects the solubility of Cu in Al. According to Hanson and Gaylor[6] there is a considerable change as a result of the separation of Mg_2Si at room temperature. The increase in hardness depends on the amount of the separated material, upon its state of dispersion and upon its coagulation. To what degree precipitation, coagulation of the precipitated particles and lattice distortions influence each other may not be determined with the aid of the hardness curves.

For a hardness maximum it is obviously necessary that there exist a favorable particle size as well as a favorable dispersion. An increase in the hardness as actually observed may not be explained by assuming a decreased particle size with the total mass of the separated material remaining constant. It is rather to be assumed according to Tammann[7] that the increase in hardness is due to an increase in size of the particles through further growth. If the speed of separation of the particles is not equal to the speed at which the particles grow in size the number of particles becomes less than the optimum, and in this way, according to W. Rosenhain[8] there occurs a decrease in hardness. In the alloys here considered this condition may be ascribed to the separation of the Mg_2Si.

W. Koester[9] has shown that the presence of nitrogen causes a separation of carbon at room temperature in the case of Thomas steel while an alloy with the same amount of carbon but with no nitrogen present did not show this separation of carbon after the same sort of heat treatments. Thus one component influences the sepa-

ration of another component. In the case here under consideration of an Al-Mg-Cu alloy the Mg (or here the Mg_2Si) causes a separation of the compound $CuAl_2$ and a corresponding increase in hardness such as would never be approached in the absence of the magnesium.

It may also be seen how in the case of the $CuAl_2$ dispersion a hardness maximum appears to be followed at once by a decrease. This very likely is to be attributed to the same causes as were previously discussed; viz. a difference in the speed of separation of the $CuAl_2$ and in the speed with which it forms crystals in the containing metal matrix. The optimum number of interference particles is at first exceeded, to be followed later by a condition in which the optimum particle size is exceeded.

In an investigation of the age-hardening of the Duralumin by means of X-rays A. Lange[10] found that during that time in which there was a rapid increase in hardness due to the separation of the Mg_2Si there also occurred a widening of the lines of the diffraction pattern of Al as well as a change in intensity of the lines and the appearance of spots not otherwise observed in the Al diffraction pattern. These spots had the appearance of being separated portions of the ordinary lines and in any case were to be found near to the usual lines. After having stood sufficiently long for the end condition of hardness to have been attained the widened lines of the Al pattern were still to be seen and in some cases weak but definite new lines were observed. Lange ascribed these new lines to the formation of a new phase according to the theories of S. Konno.[11] A. M. Portevin and P. A. Chevenard[12] showed, however, that the observed phenomena could be explained on the basis of a change in the solubility of the alloying elements and without the assumption of the formation of a new phase as demanded by the theory of Konno. According to U. Dehlinger[13] the widening of the lines might also be the result of rather periodic deformation of the crystal lattice. It is quite possible, therefore, that the phenomenon observed by Lange is due to the appearance of "ghost lines." This explanation seems to be especially probable since the atomic or molecular separations in the lattice can produce such periodic deformations.[14] Through the assembly of the original small particles to form larger particles, the deformations will partly disappear only to reappear again through further separations.

In this connection reference should be made to the results of some recent investigations by R. Hobrock.[15] Hobrock found that "ghost lines" in the diffraction patterns of *a*-iron were to be observed in layers considerably removed from the surface when an iron alloy containing 0.8% Al was nitrided in a stream of ammonia. Since at these depths there could have been no collection of the AlN molecules formed by the nitriding, the appearance of these "ghost lines" could only be attributed to periodic deformations of the iron lattice. Nearer the surface, where more nitrogen has penetrated and formed more AlN molecules, the ghost lines were not observed

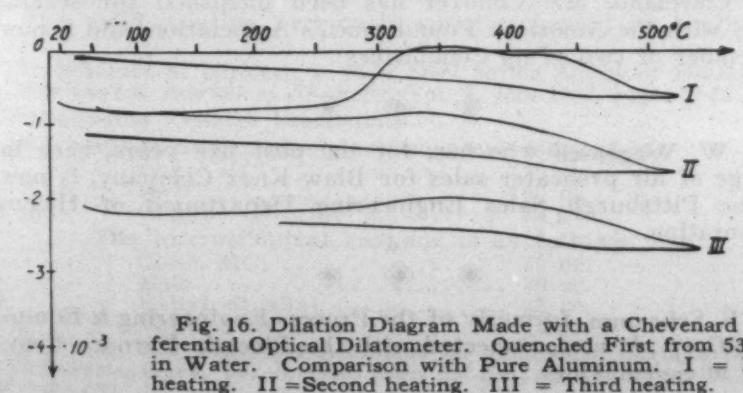

Fig. 16. Dilation Diagram Made with a Chevenard Differential Optical Dilatometer. Quenched First from 530°C. in Water. Comparison with Pure Aluminum. I = First heating. II =Second heating. III = Third heating.

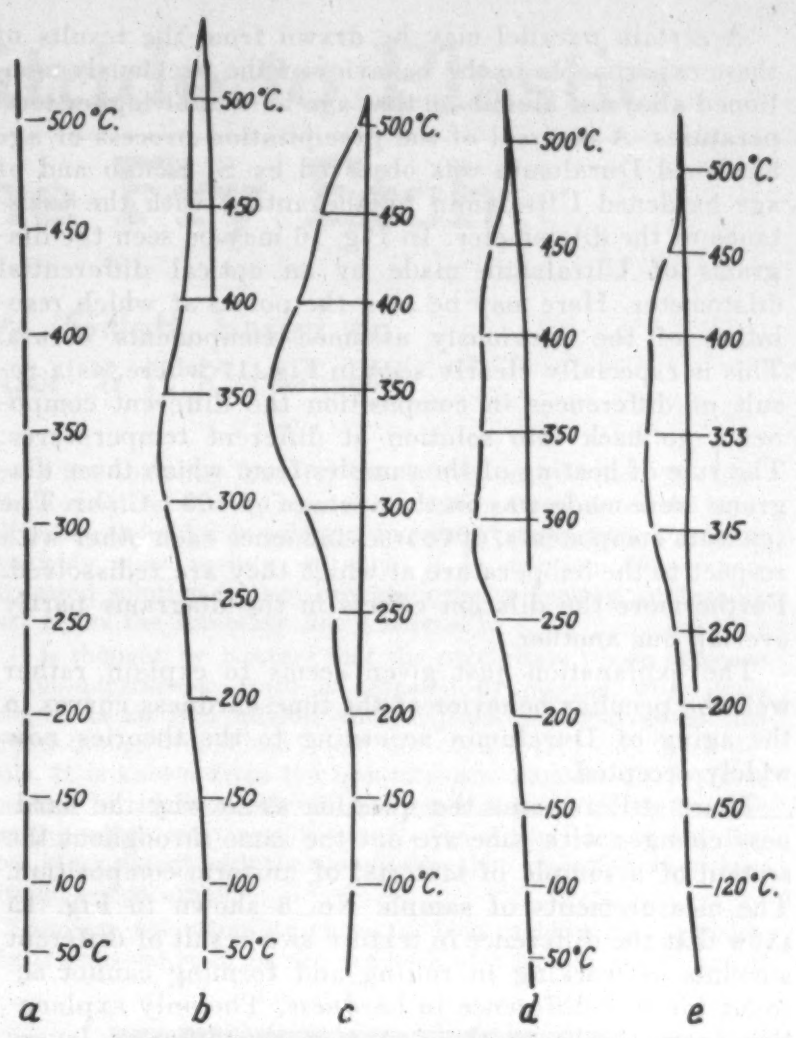

Fig. 17. Dilation Diagrams of Aluminum and Hardened Aluminum Alloys with an Improved Optical Dilatometer (F. Bollenrath). Comparison with Pure Aluminum. a = pure aluminum. b = Silumin. c = Lautal. d = Duralumin 681 ZB. e = Duralumin 681 B.

since here the larger number of AlN molecules permitted them to collect into larger particles. The ghost lines cannot be attributed to lines of the diffraction pattern of AlN since they do not correspond to any of these lines and also since the small amount of AlN present together with its low reflecting power prevented the detection of any lines of the AlN lattice in the diffraction patterns made. The greatest hardness was found where a coagulation of the AlN had occurred (nearer the surface) which seems to accord better with the hardness theory of Jeffries and Archer[16] than with that of E. Maurer.[17]

These lattice deformations cause further separation of the $CuAl_2$ in the case of Duralumin corresponding to the observations of Dehlinger. Since all of the separated molecules do not assemble to form larger particles the lattice deformations only partly disappear. Therefore in the final condition the ghost lines are still to be seen in the diffraction pattern. Also in the case of the separation of $CuAl_2$ the optimum of the number and size of the interference particles is soon exceeded so that the slip interference is reduced and the hardness is decreased.

The second increase in hardness after a long time can be explained according to Rosenhain as a result of new and further separation of particles or it may be ascribed to the formation of a new phase from the separated components or from the remaining alloying elements with the separated components. This would also afford a possible explanation of the results of the investigations of Lange.

The recent developments of F. Zwicky[18] concerning his theory of secondary lattice structure help to a large extent to explain the behavior of alloys and will undoubtedly aid materially in an explanation of the change in properties occurring during separation and coagulation of precipitating components.

A certain parallel may be drawn from the results of these experiments to the behavior of the previously mentioned alloys of aluminum that age-harden at higher temperatures. A reversal of the precipitation process of age hardened Duralumin was observed by S. Konno and of age hardened Ultralumin by the author with the assistance of the dilatometer. In Fig. 16 may be seen the diagrams of Ultralumin made by an optical differential dilatometer. Here may be seen the points at which resolution of the previously assumed components occurs. This is especially clearly seen in Fig. 17 where, as a result of differences in composition the different components go back into solution at different temperatures. The rate of heating of the samples from which these diagrams were made was on the average of 100° C./hr. The separate components, of course, influence each other with respect to the temperature at which they are redissolved. Furthermore the dilation effects in the diagrams partly overlap one another.

The explanation just given seems to explain rather well the peculiar behavior of the time-hardness curves in the aging of Duralumin according to the theories now widely accepted.

There still remains the question as to why the hardness changes with time are not the same throughout the section of a sample of material of uniform composition. The measurements of sample No. 3 shown in Fig. 15 show that the difference in texture as a result of different amounts of working in rolling and forming cannot account for this difference in hardness. The only explanation seems to lie in the fact that the different layers were cooled at different rates during the quenching. The further the layers are from the surface the slower is the conduction of heat from that layer because of the constantly decreasing temperature gradient. That the rate of cooling has a great influence upon the course of the hardening and upon the maximum end hardness may be seen from the experiments of H. C. Knerr.[19] Some of his results are as follows:

Quenching Medium	Scleroscope Hardness After 5 days		Ultimate Strength In kg./mm.² after 10 days		Elastic Limit In kg./mm.² after 10 days	
	Plates 1/16″ Thick	Plates 1/8″ Thick	Plates 1/16″ Thick	Plates 1/8″ Thick	Plates 1/16″ Thick	Plates 1/8″ Thick
Cold Water	25.7	27.2	43.4	42.8	26.8	29.6
Boiling Water	25.05	25.2	43.2	43.4	26.2	28.0
Cold Oil	24.8	26.9	43.1	43.4	27.0	28.6
Air Stream	25.0	24.1	41.7	41.2	27.0	28.0
Still Air	23.2	25.4	41.4	40.8	26.0	27.0

The speed with which the surface is cooled is not the most desirable but in the case of the previously described Duralumin samples 30 mm. thick the cooling rate of the layer at about a depth of between 0.4 and 0.8 mm. beneath the surface seems to have been the optimum cooling rate.

An explanation of this might be that with increased cooling rates the supersaturation is increased and the number of places at which separation starts is increased. This, of course, will influence the number and the size of the final particles resulting after coagulation. To arrive at the optimum number and size of interfering particles only one cooling rate is suitable.

In the outside layer the hardness does not change in the normal way. The maxima achieved are markedly lower than those in the layers beneath and occur only after longer times. This probably is the result of the extremely high rate of cooling discussed above as well as the result of differences in texture and composition which might have been affected by the heat treatments in salt and air baths.

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Colonel J. S. Ervin has been elected president and member of the board of directors of the Mackintosh-Hemphill Company, builder of rolling mills. Colonel Ervin is an engineer by training, but his entire business life has been spent in sales work since he left Ohio State University. He was recently vice-president in charge of sales for one of the largest companies engaged in the processing of steel structural materials. He was in the army during the war and is now commander of the 508th Regiment, Coast Artillery, Anti-aircraft, which is charged with responsibility for the anti-aircraft defenses in the steel and munitions-producing districts around Pittsburgh. Colonel Ervin will be surrounded by a competent technical staff. **F. Hughes Moyer** was elected vice president and senior engineering officer. He was formerly chief engineer successively of the Illinois Steel's Gary plant, the Bethlehem Steel's Johnstown plant, and the Carnegie Steel's Clairton plant. **H. E. Field**, formerly president of the Wheeling Mold and Foundry Company, and formerly vice president of the Continental Roll and Steel Foundry Company, was also elected a vice president. **F. C. T. Daniels**, an authority on rolls construction, was elected chief metallurgist and chief research engineer. Other officers elected were **Donald H. Baum**, secretary; **W. C. Rice**, treasurer; **C. Howard Paul**, assistant treasurer, and **J. R. Patterson**, divisional sales manager.

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O. D. Conover has established offices at 1740 East 12th Street, Cleveland, Ohio. He will operate in the capacity of Consulting Engineer, specializing in the development, layout and design of industrial plants, and foundries.

For the last three years, Mr. Conover has been Vice President of The Foundry Equipment Co., Cleveland, and during the preceding ten years was assistant chief engineer of The Austin Co., Cleveland. Mr. Conover has been identified for several years with the American Foundrymen's Association, and is now a member of two of its Committees.

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E. W. Wagenseil who has, for the past five years, been in charge of air preheater sales for Blaw-Knox Company, is now in the Pittsburgh Sales Engineering Department of Hagan Corporation.

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J. F. Schrumm, formerly of the Process Engineering & Equipment Corp., is now connected with the Electric Furnace Company at Salem, Ohio.

Relation of Iron-nitride to Plastic Flow and to the Fry Etch

Extended abstract by H. W. Gillett, based on paper* by C. W. MacGregor† & F. R. Hensel†

THE beginning of plastic flow occurs sharply in some mild steels, being manifested by a sharp yield point "jog" in the stress-strain diagram. In others, plastic flow develops more slowly, there is no sharply defined yield point, and the stress-strain diagram resembles that for a non-ferrous metal.

Plastic flow is manifested by the appearance of the Hartmann-Lüders lines on the surface of a low carbon steel specimen, best shown upon a polished specimen in which concentration of strain is present, as in a specimen with a hole in it. (Fig. 2). These flow lines represent non-uniform strain distribution in the material. It is often of interest to know whether or not a specimen has been subjected to plastic deformation, and how the strain has been distributed within the material.

The Fry etching reagent[1] applied to a mild steel of suitable

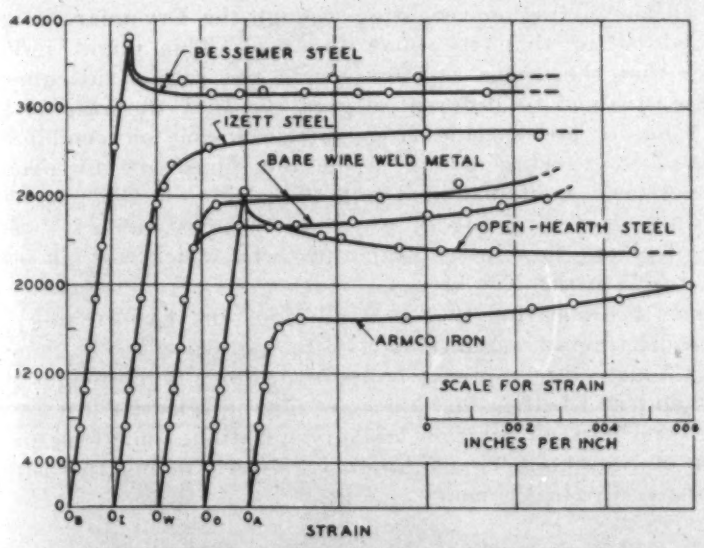

FIG 1 - PORTIONS OF STRESS-STRAIN CURVES FOR DIFFERENT MATERIALS (TESTED IN TENSION)

composition which has been strained to a suitable degree and heated to a relatively low temperature, say 200° C., reveals flow lines within the body of the steel, apparently of the same nature as the Lüders lines on the surface. It is claimed[2] that the propensity of a steel to show strain lines by the Fry etch is related to its propensity to "age" and become embrittled in boiler service. The suggested explanation[2] is that the Fry etch reveals differences in corrodibility of the more severely and less severely strained portions of the metal, and that this difference in corrodibility in the reagent is due to preferential precipitation of a hardening constituent. This is in line with the general behavior of precipitation-hardening systems, in which, as a rule, precipitation is greatly aided by previous plastic deformation, and occurs on heating (or in cases like duralumin, on standing at room temperature) which throws the precipitate out of a solid solution previously super-saturated by heating above the solubility line for the system in question. The dissolved constituent may require quenching from above the solubility line to hold it in solution, in some systems, while in others relatively slow cooling will hold it in solution. The supersaturated solution is then broken down by heating, preferably just below the solubility line ("solvus").

It is thought by Köster[2] that the precipitate whose presence in sub-microscopic form is revealed by etching with Fry's reagent is an iron nitride, and he produced evidence to this effect, though Jungbluth[3] does not corroborate him in all details. It is known from the iron-nitrogen diagram (Fry[4]) that the iron nitride-iron solubility curve has such a slope that the heating which most readily reveals the strain lines by the Fry etch is consistent with the assumption that a precipitate of iron nitride is concerned.

However, the solubility curve for iron carbide, and probably for ferrous-oxide, in ferrite are both of a similar shape, and it

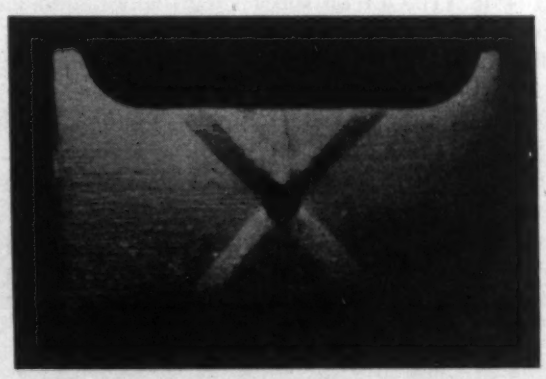

Fig. 2

is not entirely clear whether the etching effect is due solely, or primarily to nitrogen. But there is one bit of evidence that tends to indicate that nitrogen is concerned. Mild steel with increasing nitrogen up to 0.05% reveals the etch-figures with increasing ease. At higher nitrogen contents, as, for example, in bare weld metal, there is general deposition of copper from the reagent, as if the nitride were everywhere present, regardless of strain. Since there is an eutectoid above 0.05% nitrogen, this is consistent with the idea that it is nitrogen that the Fry etch reveals, and that the ease with which the Fry lines can be developed is something of a measure of the content of nitrogen thrown out of solid solution.

If this be true, then it may be possible also to trace a connection between content of nitrogen in solid solution in ferrite and the shape of the stress-strain diagram, although this side of the question was not discussed in the paper by MacGregor and Hensel. Fig. 1 shows stress-strain curves for several steels. From this it can be noted that in the samples tested, Armco iron, Izett steel, and bare wire weld metal give curves without the yield point drop, while Bessemer and open hearth show sharp drops. Inspection of Table 1 shows that the Armco and Izett and open hearth are all low in nitrogen, while Bessemer and the arc weld metals are high in nitrogen. However, the bare weld metal is so high in nitrogen that the eutectoid would be present and this might tend to "seed out" nitrogen from the solid solution. In Izett and Armco we might recall that the former is reputed to be rather heavily scavenged with alumi-

*The Influence of Nitrogen in Mild Steel on the Ability of Developing Flow Layers. *Journal of Rheology*, Vol. 3, Jan. 1932, pages 37-52.

†Westinghouse Research Laboratories.

[1] The composition of the macroscopical reagent is:

Conc. HCl	120 cc.
H_2O	100 cc.
Copper chloride	90 gr.

The microscopical reagent is as follows:

Conc. HCl	40 cc.
H_2O	30 cc.
Ethyl alcohol	25 cc.
Copper chloride	5 gr.

[2] W. Köster. Zur Frage des Stickstoffs im technischen Eisen. *Archiv für Eisenhüttenwesen*, Vol. 3, 1930, pages 637-658; see also discussion by Fry, page 657.

[3] H. Jungbluth. Alterungskergzähigkeit und Kraftwirkungsfiguren Aetzung. *Krupp'sche Monatshefte*, May 1931, pages 106-112.

[4] E. W. Ehn. Seasoning of Steel—Effect of Carbon, Nitrogen, Copper in Small Amounts. *Metal Progress*, Vol. 20, Sept. 1931, pages 59-64, 108.

Table I. Chemical and Gas Analysis of Materials Tested

Kind of Material	C	Mn	P	S	Si	Al	Ni	Cu	Nitrogen	Oxygen
Armco iron	0.038	0.033	0.008	0.015	0.011	.....	.....	.....	0.004	
Izett No. 1	0.222	0.57	0.043	0.041	0.042	0.041	0.065	.....	0.0045	
Izett No. 2	0.255	0.57	0.032	0.038	0.040	0.041	0.072	.....	0.0053	
Open-hearth No. 4	0.145	0.45	0.024	0.062	0.038	0.002	.....	.....	0.0042	
Bessemer No. 1	0.140	0.94	0.114	0.145	0.018	0.002	.....	.....	0.011	
Oxyacetylene weld metal	0.033	0.14	0.027	0.021	0.003	.....	.....	.....	0.013 0.024	0.054
West. weld metal	0.070	0.42	0.012	0.018	0.19	.....	.....	.....	0.05	0.09
Bare wire arc-dep. weld metal	0.08	0.29	0.018	0.033	0.009	.....	.....	.....	0.124	0.21

num, Armco probably less heavily, so that one might expect that a good deal of the nitrogen would be bound as aluminum nitride, which is presumably far less soluble in ferrite than iron nitride, so that the iron-nitride content, could it be determined, would presumably increase in the steels in the order, Izett, Armco, open hearti Bessemer. This would place them in the order of ease of Fry etching. (Bessemer etches clearly in a few minutes, open hearth may require several hours and even then may not etch clearly, while Armco and Izett will not etch.) It is also consistent with the shape of the stress-strain diagrams if it be assumed that the diagrams reflect the presence of iron-nitride in solution.

Unfortunately, the scavenging elements that may be used to bind nitrogen into some other less soluble nitride than that of iron, also bind oxygen and further evidence is needed to clarify the problem. However, an assumption that the shape of the stress-strain curve is connected solely with the nitrogen content is probably untenable. Dr. Mehl[5] states that several thousand tensile curves of Armco iron show very pronounced yield points even though the nitrogen averages only about 0.003%. The yield point can be made to come and go, though not all the variables controlling its presence or absence are as yet clearly understood. The presence or absence of nitrogen does not appear to be the controlling factor in the case of Armco. This, of course, does not prove that the presence of larger amounts of nitrogen may not be a causative factor of great influence. According to a footnote in MacGregor and Hensel's article, picric acid in alcohol showed strain lines in some batches of Armco iron and not in others.

In work at Battelle Memorial Institute samples of Izett and of Armco have shown pronounced jogs in the stress-strain curves, but, comparing these with Bessemer, for example, the extent of yielding under constant load, at a slow rate of stretching, is different in each case.

Izett steel does not reveal the strain lines within the metal by the Fry etch, although the distribution of strain is doubtless much alike in all the steels, as is indicated by Izett's showing Lüders lines on a polished surface. The distribution of strain might, however, be altered by the different rate of slip in a metal that does not show a sharp yield point.

For all these reasons, it is of interest to know more about the relation of nitrogen to the Fry etch, and to find whether and by what means Izett and Armco can be made to reveal flow within the metal. In the paper under consideration, MacGregor and Hensel show that an oxy-acetylene weld metal with 0.02% nitrogen etches readily. Arc-weld metal deposited with coated electrodes and containing 0.05% nitrogen etches very clearly. Bessemer of 0.011% nitrogen etches very clearly so that any mild steel with from 0.01 to 0.05% nitrogen should be readily amenable to the Fry etch.

By light nitriding of a polished surface of open hearth steel or Izett steel to produce a suitable nitrogen content and then straining, all these materials, which without nitriding either etch slowly or do not etch at all, showed clear lines under the Fry etch. This corroborates Köster's contention that the Fry etch reveals iron nitride. It would be of interest to try to raise the iron-oxide content of similar specimens, etch and see whether any increase in etchability were obtained.

Another interesting observation was that Izett, strained before nitriding, and then nitrided, revealed strain lines on etching without a further reheating, though the Fry micro reagent worked better than the macro reagent. This might indicate either that there was a difference in the rate of nitriding of ferrite strained to different degrees or that the rate of precipitation of the nitride from solid solution on cooling was affected by varying strain. While this phase of the problem has not been examined in detail, the utility of the process of nitriding after straining in revealing the existence of a beginning of plastic flow in a strained material which will not otherwise react to the Fry etch, is evident.

Since the reheating that best develops the Fry etching lines is a short time at a relatively low temperature, say ½ hour at 200° C., instead of a higher temperature or a much longer time, one wonders whether the etch may not be selective for nitride in a form between solution and precipitation, something of the order of the "knots" postulated by Merica in his Institute of Metals lecture in February.

[5] R. F. Mehl. Personal communication, Jan. 29, 1932.

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A. S. T. M. - A. S. M. E. Joint High Temperature Committee Research

Although funds for research activities to be carried on either within or without the laboratories of industrial firms have not recently been as readily available as they were in more prosperous times, it is encouraging to note that some projects of extreme importance can still be financed.

The need for accurate data on the behavior of the 18-8 chromium nickel type of alloy at high temperatures is so pressing that the Joint Research Committee of the A.S.T.M. and A.S.M.E. on Effect of Temperature on the Properties of Metals has outlined a comprehensive study of the effect of carbon content and of preliminary heat-treatment on the chemical, mechanical and structural stability of this type of steel, both in wrought form and as castings.

Much of this work has been underway in the laboratories of the organizations represented on the Committee, but some parts of the work which require long-continued tests demanded special attention if the results are to be obtained with reasonable promptness. Such phases are the high temperature endurance properties and the resistance to creep. Preliminary evidence indicates that up to somewhere around 1000° F. the endurance properties are important, but that at still higher temperature the limiting stress is that under a fixed load, i.e., the creep resistance. More is known about these phenomena in pearlitic steels than in austenitic steels; hence, their study is of fundamental interest as well as of importance in regard to immediate commercial problems in the industrial application of the austenitic steels.

The Joint Committee set out to secure $20,000 for a three year research program, and while only about half of this sum has so far been secured, the Committee and the groups that have so far appropriated funds for the experimental program have, however, felt that the work should be started without delay.

These sponsoring groups are the Engineering Foundation, the National Electric Light Association, the American Petroleum Institute, the National Research Council and a group of alloy castings manufacturers through the Corrosion Resistant Alloy Founders' Division of the Steel Founders' Society of America, Inc. Further financing of the work by those to whom the results will be valuable is hoped for so that the experimental program may go on without interruption.

The Committee's program on endurance at high temperatures is being carried out at the University of Illinois, and that on creep properties at Battelle Memorial Institute. Active work began at both institutions on March 1st.

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A volume of abstracts useful to steel foundrymen has been published under the title "A Resume of Published Data on Steel Foundry Practice," as Circular No. 25 of the Ohio State University Engineering Experiment Station. The work was compiled by A. H. Dierker, research engineer of the station, and H. B. Kinnear, consulting engineer.